EVALUATION OF METALS AND ALLOYS FOR USE AT TEMPERATURES EXCEEDING 1000 F.

BY

C. J. FLESSNER

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by

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Lieutenant, U. S. Navy

June 1950

CARNEGIE INSTITUTE OF TECHNOLOGY

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A GAS TURBINE POWER PLANT

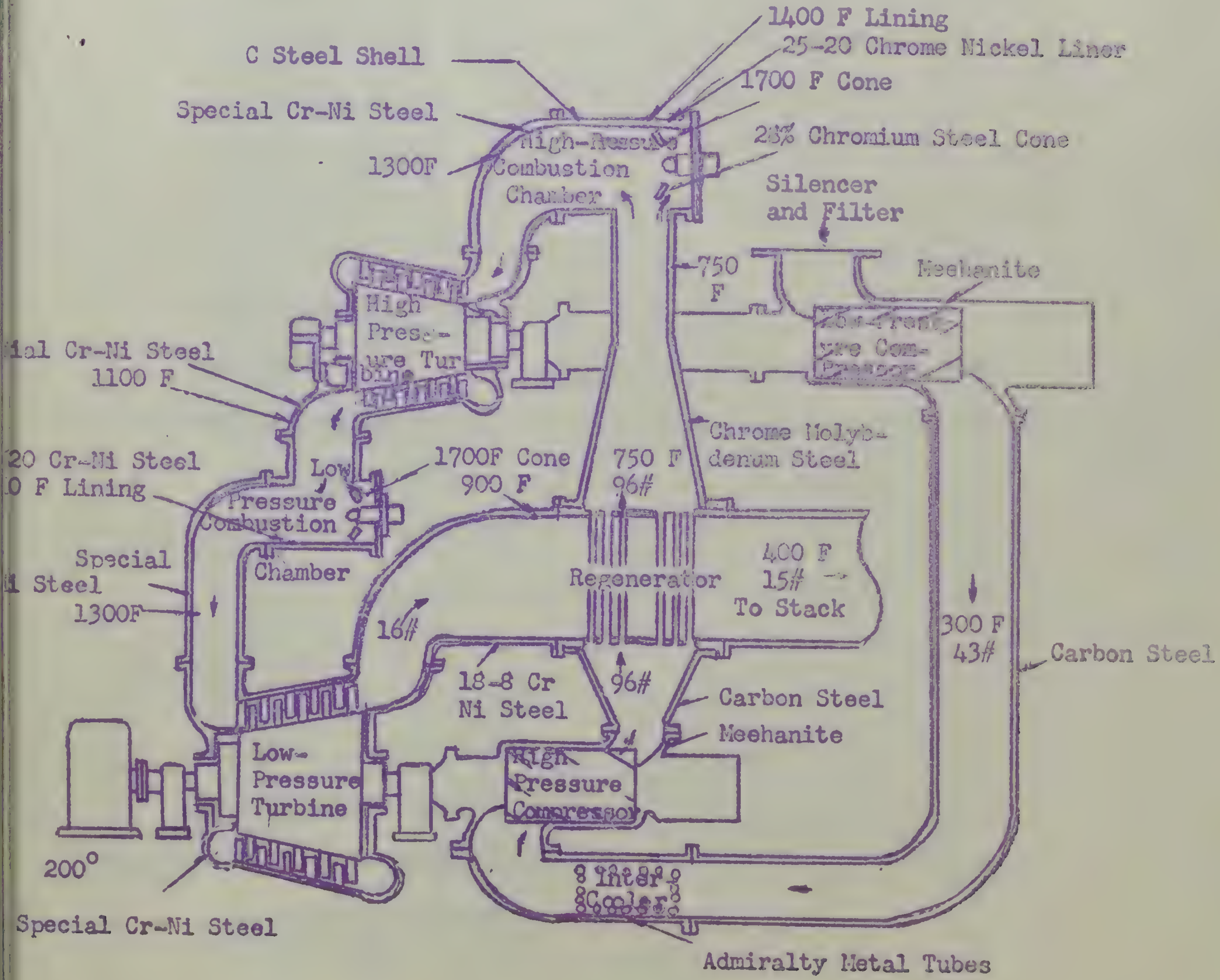


Figure 1. Gas turbine power plant. (24)

PREFACE

There is much information and data published on metals and alloys for use at temperatures exceeding 1000°F. Unfortunately much of the data is presented in a factual manner with a minimum of interpretation and correlation. From such a mass and maze of compositions, temperatures, times, tests, methods and many other variables it is difficult to get a clear picture of just what is desired in a metal or alloy for use at high temperatures.

There are several reasons for this state of affairs. Only comparatively recently have steels, for example, been sold primarily on the basis of mechanical properties rather than on the basis of chemical compositions.

High temperature alloys were needed in a hurry for gas turbines, jet engines, turbosuperchargers, and other military applications during the recent war. Time simply was not available to make fundamental studies of phenomena occurring at elevated temperatures. As a result many varied and diverse alloys were manufactured. The successful alloys were so deemed on the basis of final performance in the application which is quite logical, but there was not a standard system of laboratory testing on which one could rely to rate the various alloys prior to the service test. Thus the results were highly empirical. One satisfactory alloy

might have three or four alloying elements while another might have ten alloying elements.

In order to arrive at some body of general information and considerations which might be applied to selection of metals for use at elevated temperatures, an attempt has been made in the following report to classify existing knowledge under various mechanical and physical properties.

INTRODUCTION AND SCOPE

In 1948, a report by Scarborough (1) traced the development of heat resistant materials. The coverage given by him can best be summed up by a duplication of the table of contents of his report.

I. Introduction

II. Early Elevated Temperature Work
    origin
    corrosion and oxidation
    intergranular corrosion
    high temperature strength and ductility
    fracture
    alloys and alloying element effects

III. Evaluation of Elevated Temperature Material
    data and correlation
    creep

IV. Recent High T emperature Work (1939 - 1948)
    new fields
    iron-base alloys
    nickel-base alloys
    cobalt-base alloys
    chromium-base alloys
    Cr - Ni - Co - Fe alloys
    comparisons and properties of super alloys

V. Future of Heat Resistant Materials

Some of the material in the present report overlaps to a certain extent the material of the earlier report by Scarborough, but overall the organization and emphasis is radically different. In line with the more modern approach of studying steels for example by classifying them according to structure and of specifying

steels on the basis of mechanical properties and hardenability, as little reference will be made to specific chemical compositions and individual alloying elements as is consistent with understanding the remarks herein. In other words, more importance is ascribed to the properties of the particular alloy than to the composition since the alloying elements are only one of the many variables such as melting, ingot practice or casting, working, and heat treatment which determine the mechanical properties of the alloy. Of course there are exceptions to the above where an individual element is very important such as the necessity of having chromium present in sufficient amount for corrosion resistance, graphitization as a result of an excess of silicon with sufficient carbon, and the non-protective nature of the oxide of molybdenum. These exceptions with several others are discussed in the appropriate sections of this report.

While on the subject of the effect of specific alloying elements in heat resisting steels, it might be remarked that Bain (2) has presented the general subject of alloying elements in steel and Dobkin (3) has described the effects of alloying elements in heat resisting steels.

W. O. Binder (19) has classified high temperature alloys in the following manner:

A. Work-hardened alloys

  1. Up to 70% Fe
  2. Cold worked austenitic structure
  3. High yield strength
  4. Limited to 1350°F. service temperature

B. Heat-treated alloys

  1. Non-titanium bearing

     a. Up to 30% Fe
     b. Face centered cubic lattice structure
     c. Lower yield strength than class "A"
     d. Subject to change in solid solubility
     e. Greater high temperature strength than class "A"

  2. Titanium bearing

     a. Up to 55% Fe or not less than 6% iron with balance nickel and cobalt or nickel
     b. Face centered cubic lattice
     c. Characterized by high yield strength
     d. Limited to about 1350° service temperature

C. Cast alloys

  1. Cobalt-base types

     a. Contain up to 1% iron
     b. Face-centered cubic lattice
     c. Moderately high yield strength
     d. Subject to solid solubility changes
     e. Greater high temperature strength than class "B"

  2. Chromium - base alloys

     a. Contain up to 25 percent iron
     b. Body-Centered cubic lattice
     c. Extremely brittle at room temperature
     d. Possess high strength at elevated temperatures

The above classification is not the only classification given to high temperature alloys in the last few years. It is probably not the best or most complete outline. But there are

several interesting points to note on it. The alloys have been grouped as cast or wrought, heat treated or work-hardened, and reference is made to lattice structure, solid solubility, yield strength, high temperature strength, and ductility. In the earlier days of high temperature metallurgy, just a few years ago, it was believed that high temperature strength could be increased more expeditiously by varying composition and keeping processing variables constant (19). This resulted in a large number of complex composition which defy systematic classification. Yet it was noted that most of the alloys were subject to solid solubility changes at 1200 - 1800°F. It is logical to assume then that these alloys are similar in that they probably get their strength from the alloying elements by the same mechanism.

In order to properly evaluate an alloy for an engineering application one must know the conditions of operation of the piece of equipment. For example the selection of material for the inner wall of a cooled, liquid propellant rocket motor (16) would be on a different basis than the selection of material for blades of a gas turbine (23), (24), (29), (30). The gas turbine and turbo-supercharger are very much alike except in one respect - size. Thum (41) has said: "It is one thing to build and blade a small disk for a turbo-supercharger which operates at say 1500°F. maximum but at fairly low centrifugal stress, and another thing to construct and mount a similar blade, which may be fifty times as heavy to run at peripheral speeds exceeding the velocity of sound. This

transition is so difficult that the gas turbine, so far, is designed to run 200 or 300 degrees F. cooler than the turbo-supercharger despite the fact that thermodynamic efficiency increases faster than temperature. Time of operation must also be considered. Is the turbine expected to run a maximum of 1000 hours or is the expected life to be 100,000 hours? Is operation to be steady such as a power house application or is operation to be cyclic such as in a transportation application?

In order to restrict somewhat the many faceted problem of evaluating high temperature alloys, the bulk of this report will be concerned primarily with properties of interest with respect to material for gas turbine rotors and blades. The rotor and blades of a gas turbine are perhaps the most difficult application to meet since the temperature of the blades is high, the stress system is complex to the point of being insoluble, and just about all of the properties and tests associated with high temperature alloys must receive consideration. A schematic diagram of a gas turbine photograph is shown in Figure 1.

Originally, it was planned to describe testing methods and equipment in some detail. But due to limitations of time, it was decided to make rather brief limited references to testing equipment. Complete descriptions of the equipment mentioned may be found in the articles referred to in the bibliography.

Evans (30) has presented the points which must be considered in evaluating materials for gas turbine service in a unique manner:

Temperature

Oxidation, corrosion

Thermal expansion

Thermal conductivity

Effect of fluctuations in exposure

Stress - to - rupture

Creep

Creep relaxation

Unknown factors

train
(allowable
deformation)

Time
(expected service life)

Effect of fluctuations

Effect of concentrations

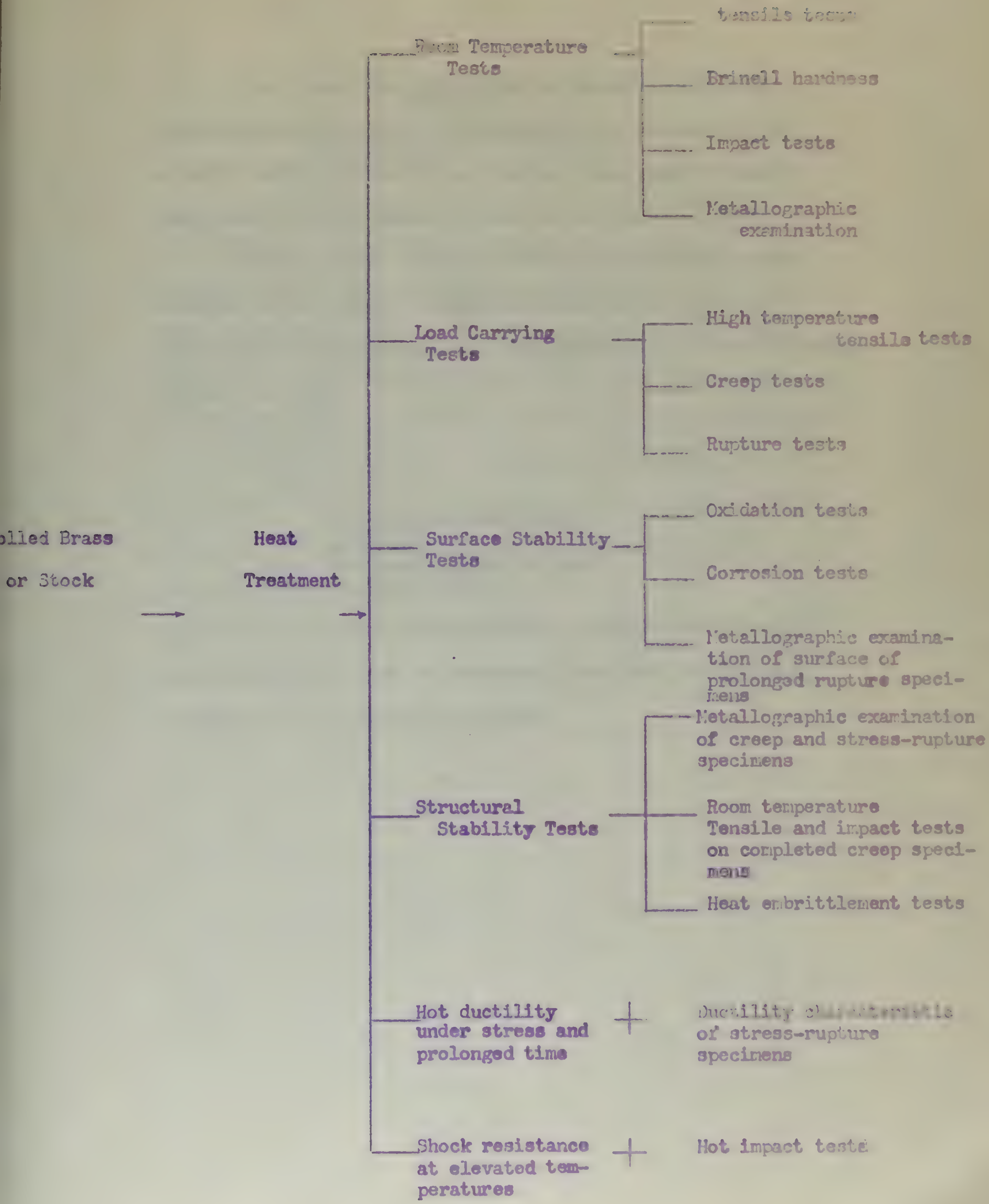
Fatigue

Damping capacity

Stress

According to Evans, the design engineer first specifies the limiting conditions of stress, strain, time, and temperature. Then the metallurgist selects the proper material considering the factors listed inside the box, always with the four limiting conditions of stress, strain, time, and temperature in mind.

The Timken Roller Bearing Company (134) employs the following tests depicted in the diagrammatic sketch below in determining the suitability of steels for high temperature service.


olled Brass
or Stock
Heat
Treatment
Room Temperature Tests
tensile tests
Brinell hardness
Impact tests
Metallographic examination
Load Carrying Tests
High temperature tensile tests
Creep tests
Rupture tests
Surface Stability Tests
Oxidation tests
Corrosion tests
Metallographic examination of surface of prolonged rupture specimens
Structural Stability Tests
Metallographic examination of creep and stress-rupture specimens
Room temperature Tensile and impact tests on completed creep specimens
Heat embrittlement tests
Hot ductility under stress and prolonged time
Ductility characteristic of stress-rupture specimens
Shock resistance at elevated temperatures
Hot impact tests

The reason for presenting the classifications of high temperature alloys, of properties to be considered, and of tests usually performed has been to illustrate the basis upon which the major points of this report were selected.

The table of contents serves as an outline for the material of this paper. The items listed under each major topic are incorporated into the subject matter with the following emphasis in mind:

For example:

Creep

A. Theory or present status
B. Application
C. Current experimental work
D. General remarks

While the topics have been physically separated for purposes of classification, it must be remembered that they are dependent on each other to a very large degree.

CHAPTER I

STRENGTH, TOUGHNESS, DUCTILITY

In reviewing the literature of high temperature metallurgy, the reader very frequently encounters the statement that strength, toughness, and ductility are desirable in alloys for use at high temperature (48), (68), (76). Strength and toughness are rather obscure terms; no one can really give a binding definition for either of them. Ductility is usually associated with elongation or reduction in area. Toughness is usually associated with notch sensitivity. Very often, toughness and ductility are loosely referred to as a single property or characteristic.

Equi-cohesive temperature (136) is associated with strength and temperature (34), (60), (63). This temperature is defined as the temperature above which the material fails with a relatively brittle fracture which follows grain boundaries or (phase boundaries), and below which failure occurs as a relatively ductile process in a transcrystalline manner. Another way of stating the definition is that it is the temperature above which a material fails by cleavage due to strain rate, rather than by slip. Some people have objected to the use of the term equi-cohesive temperature on several grounds (63) while other investigators have used the term in a purely descriptive manner. The principal objection is that the time of application of stress determines whether the fracture is brittle or ductile. Thus the equi-cohesive temperature is not a true constant. This subject is developed further in the sections on creep, stress rupture, and hot tensile testing.

Some very fundamental studies have been made by Osipov recently on the principles of alloying of hot strength metallic alloys (35), the relation between the melting temperature and the high temperature strength of alloys (36), and factors in the high temperature strength of heterogeneous alloys(37). Osipov has considered the general aspect of the hot strength of homogenous solid solutions having B. C. C. and F. C. C. structures on the basis of the $N(\varepsilon)$ curves for iron. $N(\varepsilon)$, the density of electron states, in α and γ irons is plotted versus , the energy

of the electrons in Rydberg units. Electrons must change their energy and distribution in the crystal structure when the metal is plastically deformed. The probability of redistribution is lower (therefore the resistance to deformation is greater) the fewer are the permitted states of electrons in the upper occupied levels. Osipov then states the following principle: "A homogeneous solid solution with a minimum density of electron states in the region of the higher occupied levels of the N ($\mathcal{E}$) curve must possess the maximum heat resistance." The addition of a B. C. C. element to a F. C. C. solvent results in greater hardness and strength at high temperatures as a result of non-uniformity of the electron distribution. Metals with different crystal structures and widely separated in the periodic table tend to form intermetallic compounds. The electron distribution in such metals also differs sharply, so an increase in strength of a F. C. C. solid solution may be expected when several alloying elements that form intermetallic compounds among themselves are added.

Osipov explains the relation between melting temperature and high temperature strength in this manner. Near the temperature of crystallization, the liquid and solid have much in common in the distribution of atoms, in the character of thermal motion, and in the forces of interatomic interaction. Experimental data on iron-chromium, iron-nickel, cobalt-nickel, and manganese-nickel

alloys confirm the above views to a certain extent. The mechanical strength of the above alloys was evaluated at 1100°C by the method of Kornilov. In this method, the alloy showing the least deflection under the action of centrifugal force has the best high temperature strength.

In another article (37) Osipov attempts to explain the reason why the high temperature strength of some heterogeneous alloys is less than the strength of the single-phase alloy, while in other systems, the reverse is true. The transition zone that exists between the grains of a single-phase alloy as compared with the zone between the two different phases in a heterogeneous alloy appears to be the factor of main importance. The mechanism of flow is assumed to be the same as that of diffusion. If the melting point of the transition zone is low, diffusion of this grain boundary material into adjoining grains occurs rapidly. Eutectic systems have lower melting temperatures than the pure phases. Therefore the high temperature strength is lower for heterogeneous alloys in eutectic systems than for the single phase alloy.

The remarks of Osipov, stated above, are based on rather meager experimental evidence but do indicate an approach to the subject of creep strength and rupture strength, for example, which is necessary to arrive at some conclusions that may be relatively free of anomalies.

Ductility has been a controversial subject in the field of high temperature alloys, and recently greater emphasis has been placed upon the percent elongation caused by applied stresses over varying time periods (11). In cyclic operations of machinery (heating - cooling - heating) if plastic flow from thermal stresses occurs on each high temperature phase of a cycle, high stresses may result each time the minimum temperature is reached. In situations like this, many engineers will specify room temperature ductility as the major or primary requirement relying upon the ability of the material in changing its size to redistribute and reduce the thermally induced stresses (48). In many cases a minimum elongation of 5% in 2000 hour stress rupture test is considered desirable (11). Cross and Simmons (47) have said: " However, low room-temperature ductility does not necessarily mean inadequate ductility at the elevated temperature of operation, and the millions of these cast buckets that gave satisfactory service are sufficient proof that the final answer on ductility is in an actual service test and not in an arbitrarily chosen standard of ductility in the laboratory." Cross and Simmons are referring to vitallium buckets in turbosuperchargers in the statement above. They are not objecting to ductility, per se, but to the method of evaluating it, since the fact that the vitallium turbosupercharger blades performed satisfactorily under cyclic operating conditions would indicate that local elongation occurred

during periods of overload and sufficient relief of stress concentration as a result of cooling was accomplished (11) to prevent failure. Siegfried (127) favors describing the behavior of a steel under stress at elevated temperatures by two strength values -- the resistance of the material to deformation and the resistance of the material to fracture. Resistance to deformation is defined as the limiting stress under which a cessation of creep eventually takes place. Resistance to fracture is represented by the load resulting in deformationless fracture after an infinite length of time. Resistance to deformation might be determined by a relatively short time test provided that the internal structure of the interior of the test specimen would not change during extension of time beyond that at which limiting stress occurs and provided further that the surface of the test specimen remained unaffected by ambient conditions. With the above requirements in mind, which incidentally Siegfried did not consider, Siegfried suggests a combination short time, long time test on notched bar specimens in tension in which the resistance to deformation and the resistance to fracture are obtained.

CHAPTER II

CREEP AND STRESS RUPTURE

CREEP

There really is no generally accepted theory on the mechanism of creep. Andrade, Kanter, Seitz, Orowan and others have advanced theories, but for the most part these are in turn dependent on other theories and assumptions which are not universally accepted. Such ideas as viscous flow at grain boundaries and within grains(Andrade 137), viscous flow as a diffusion process (Kanter 138), dislocation theory applied to transient creep (Seitz and others 144), and micro creep in some crystals imposing an elastic stress upon other crystals (Orowan 139) have been suggested. It is not the intent to discuss these theories for the reason offered above.

Interpretation and presentation of creep data has appeared in several variations. McVetty and Nadai (141) have introduced a method which involves the establishment of a hyperbolic sine relationship between the minimum creep rate and stress over a range of temperatures. By cross plotting, a similar relationship is established between the minimum creep rate and temperature over a range of stresses. A similar procedure is followed with respect to the deformation obtained during the first stage of creep. From a combination of the above data curves

Creep Curves

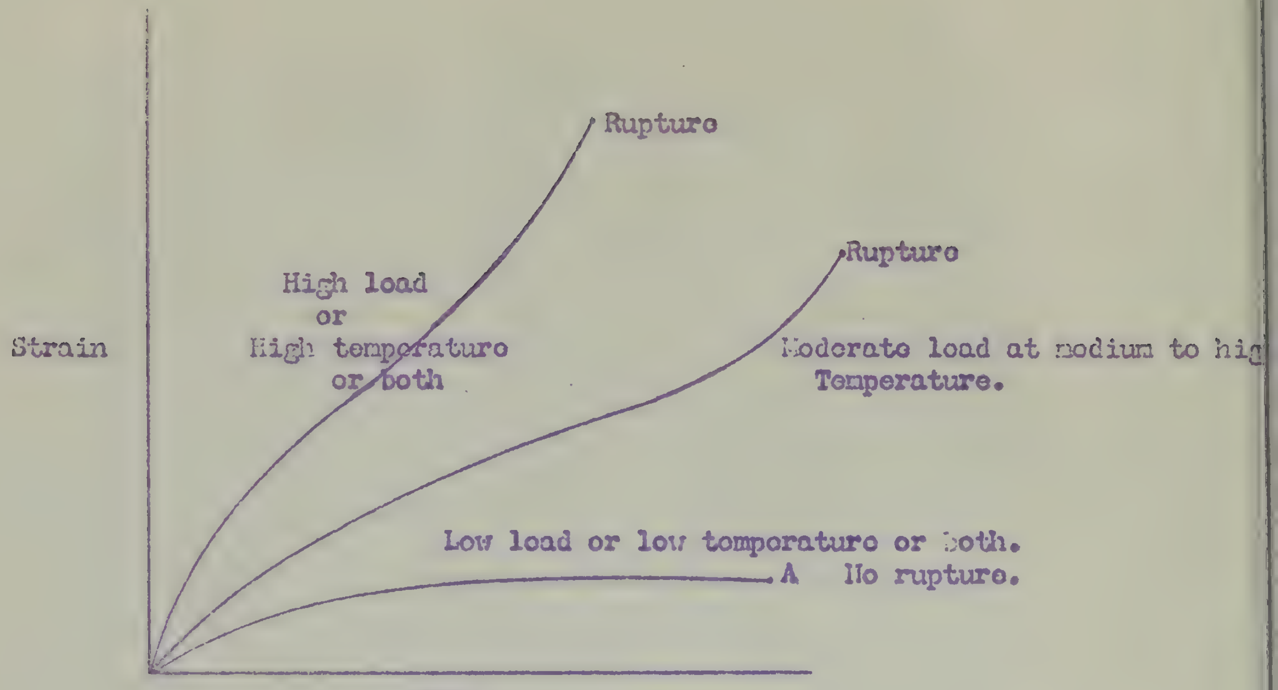


Figure 2.

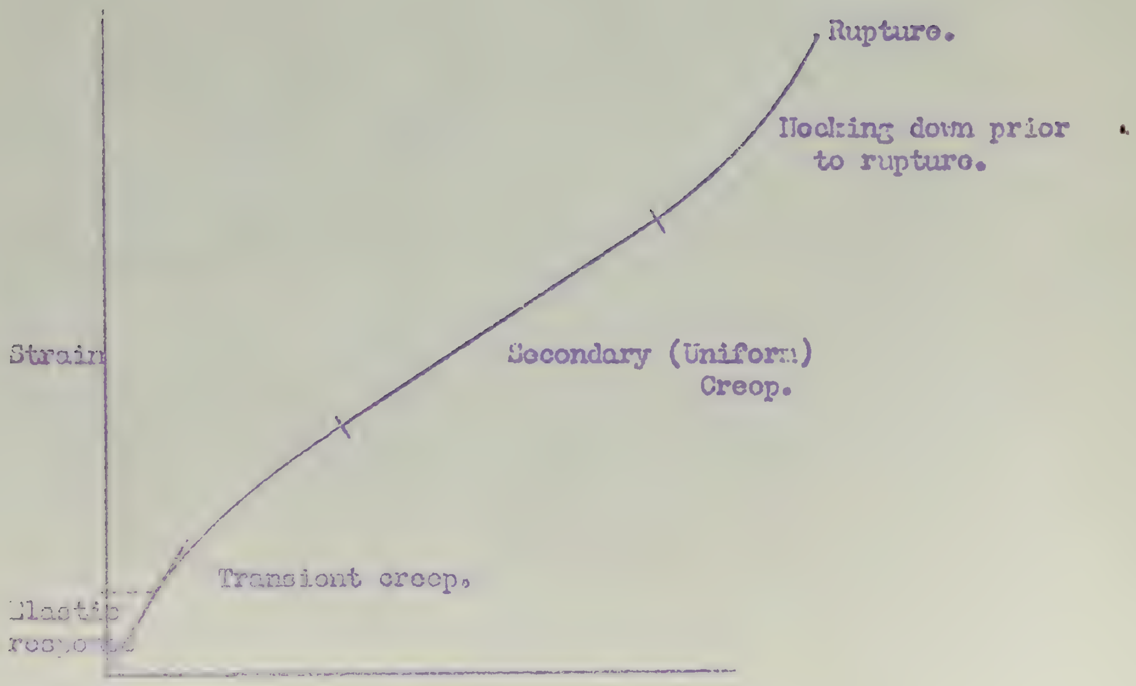


Figure 3.

of total deformation versus time over a range of stresses at constant temperature are plotted. Sets of curves may be drawn over the entire range of temperatures used. For a complete discussion of the application of this method see reference (142). This particular method has somewhat limited value since application of the above chart is restricted to tests at exactly the same temperature. Even slight variations in temperature result in a marked departure from these curves.

Creep data are usually plotted on curves as shown in Figures 2 and 3. The creep curve thus consists of three sections; transient creep, secondary creep, and third stage creep just prior to rupture. In general the greatest interest is in the section of the curve known as secondary or uniform creep. According to ASTM designation E22 - 41 results of the creep test may be plotted in the following manner: (145)

> (a) "Extension-time curves should be plotted on ordinary rectangular coordinate paper at amply large scale, and such curves should state whether or not elastic extension is included and, if so, what it is. Time is preferably expressed in hours."
>
> (b) For a particular temperature, log-log plots of stress versus rate of creep are desirable. Such plots are useful within the limits of determined values.
>
> (c) For each material a plot of log strength versus temperature is desirable. For the higher temperatures, such plots approach straight lines although the plotting of data on semilog charts often results in a straight line relationship, extrapolation of such lines has been found to be unjustified."

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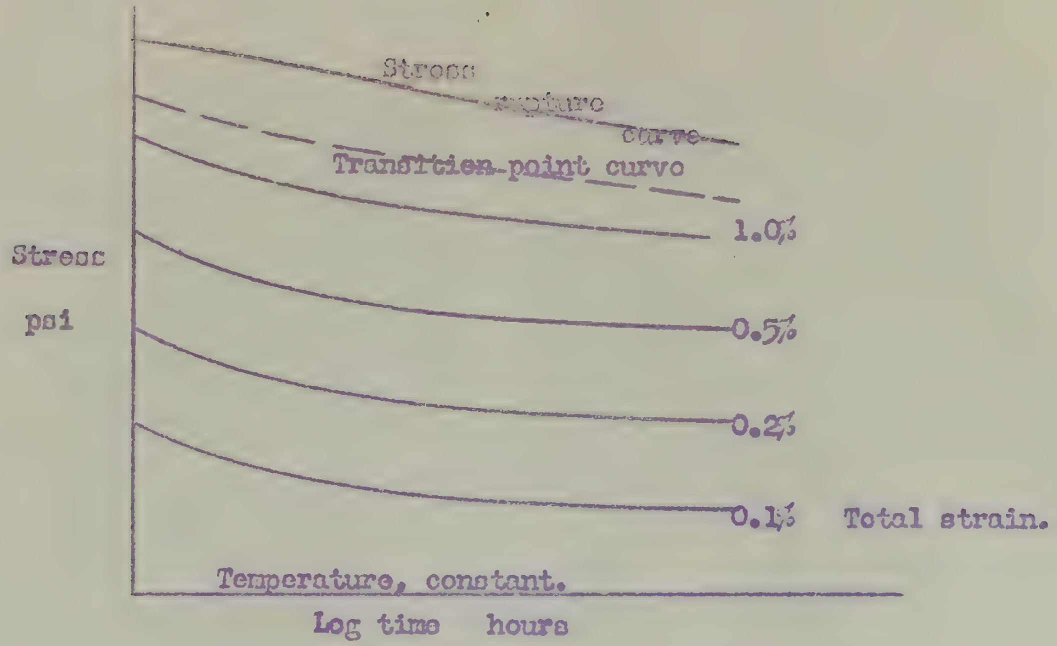


Figure 4 Curves for designers. (30).

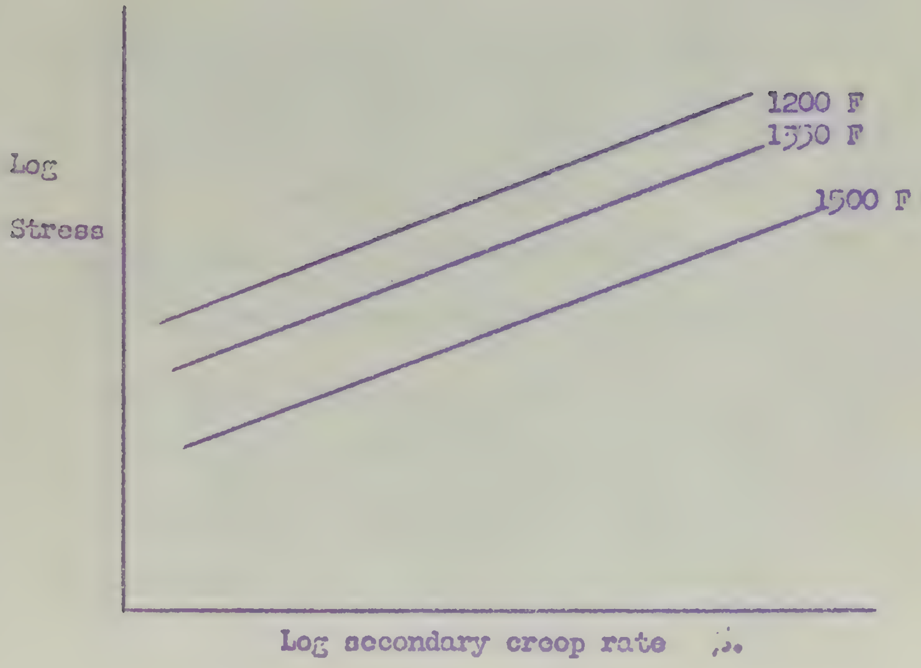


Figure 5. (30)

Creep data has been the successful basis of design for steam turbines to operate at temperatures up to 950°F with an expected life of at least 10 years (30). Particularly where close tolerances must be maintained, design is based on creep strength. For example, steam turbine parts are designed on the basis of 1% elongation in 100,000 hours (11). The original Elliott gas turbine was designed for a life of 10 years and for stresses in the neighborhood of 6-8000 psi (24).

Although it has been mentioned that the main interest is in the second stage or uniform creep part of the creep curve the initial amount of deformation from transient creep must be considered in design calculations. Creep curves for various temperatures are generally parallel. Two groups of curves each with parallel slopes may exist - one group for temperatures above and the other group for temperatures below the recrystallization temperature.

When sufficient creep and stress to rupture data exist, so called design curves are plotted (108)(30). The design curves as illustrated in Figure 4 give stress for a given temperature that will produce a certain deformation in a given time. Common practice for designers is to design with a stress which will result in a secondary creep rate of .001% per hour for 1000 hours or a factor of safety of 50% is applied to the stress which will produce a creep rate of 0.01% per hour for 1000 hours. Quite often the extension in the transient stage of creep is approximately equivalent to the extension in the secondary stage of creep for 1000 hours(30).

The rules of thumb above are strictly empirical. The "transition point curve" is plotted from the stresses and corresponding times when the creep rate increases to 10% above the secondary rate. It is important that the designer choose or design the structure for stresses which will not cause extensive elongation for time periods over 1000 hours since very little data has been obtained about the ability of the various metals and alloys to deform for long periods of time without premature fracture. For this reason, the designer should probably stay well clear of the transition point curve and limit stresses to the .1% curve if he expects the structure to last ten years (30). In 1942, the General Electric Company completed a series of 100,000 hour creep tests at 842°F on SAE 4330*. To illustrate order of magnitude figures:

1. 25,000 psi stress caused third stage creep after 70,000 hours.
2. 21,000, 17,000 and 13,000 psi caused constant or gradually reducing secondary creep rate for full 100,000 hours (30).

842° is considerably below 1000°F, the minimum temperature in which the main interest of this paper lies. But it is interesting to compare some of the above figures with data obtained on much higher alloyed steels or more complex alloys at higher temperatures. For example, Grant (60) in working on the development of alloys suitable for long - time operation in gas turbines to be operated at 1350°F

* See appendix I for compositions

25 Cr-20 Ni (Type 310)

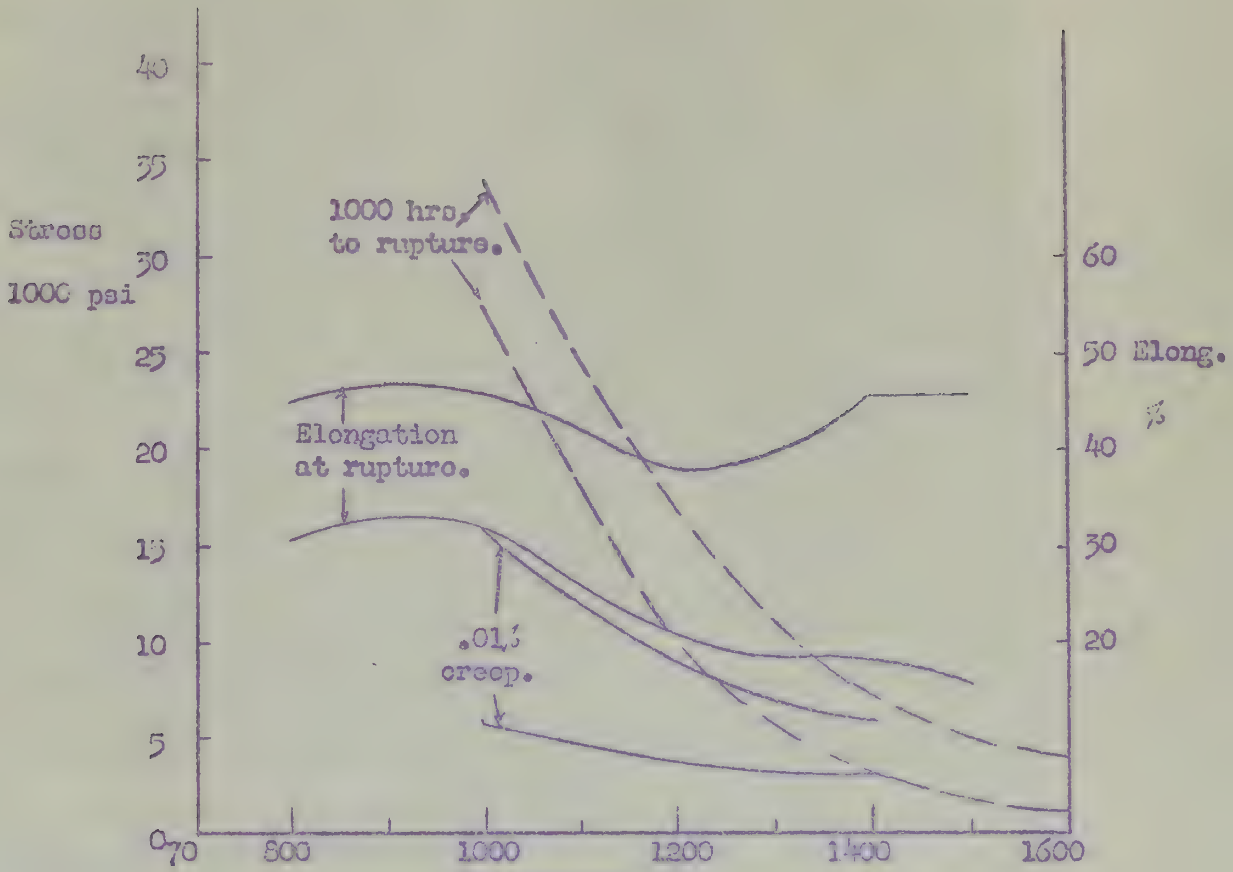


Figure 6. (30).

High temperature design information. Spread is due to difference in heat treatment . High creep and rupture values above 1000F are from material water quenched from 2150F. The low values are for material air cooled from 1700F.

to 1500° F set up a goal of a minimum creep rate of 0.00001% per hour (1% in 100,000 hours) at a stress of 7,000 psi and a temperature of 1500°F.

Another convenient way of plotting creep data is to plot log stress versus log of the secondary creep rate at various temperatures. These data plot as straight lines, as indicated in Figure 5. Such straight lines encourage interpolation and extrapolation, but there are some dangers in doing this. Creep curves which have been conducted for the purpose of obtaining a rate of deformation tell nothing about the time it will take for a metal to fracture under the loads applied, and nothing about when to expect failure under lower loads (64). The reason for the above statement and the empirical statements mentioned regarding proper selection of a safe rate is further explained together with the stress rupture test.

After examining data on 15 of the older high temperature alloys, Evans (30) has remarked that there may be as much as 100% spread in the stress necessary to produce secondary creep of 0.01% in 1000 hours. This applies particularly for temperatures above 1000° F. In Figure 6, an abnormal spread purposely caused by heat treatment is shown. The differences shown here between bars air cooled from 1700°F as against bars water quenched from 2150°F could be the result of water quenching small and large sections. Other factors of structure, melting practice, heat treatment, and working could also cause a spread in results. These effects are discussed in

the sections on structure, melting and casting practice, heat treatment and working.

The creep test may not be the best test to use for an engineering application on a large scale. The test takes a very long time to run. For example, 10 years is approximately 87,000 hours. Yet secondary creep rate values may be desired over a period of 100,000 hours. For gas turbine applications the strain measurements must be very accurately taken since the creep rate is very small. For example, if design is predicated on 1% total elongation in 100,000 hours - this elongation to be the result of secondary creep (uniform creep) - the extensometer or measuring device if taking daily readings would have to be accurate to the order of 5 millionths of an inch. This is so because a creep of 1% in 100,000 hours is equivalent to a creep rate of .00001% per hour or 2×10^{-6} inches per hour for a 2 inch gage length (64). During this great length of time, the temperature must be maintained very accurately over the entire gage length of the specimen (53). It has been remarked that temperature control on furnaces for creep machines should be ±1C for temperatures up to 900°C (59). For certain loads on certain materials, an increase of 10°F may double the creep rate. Expressing the creep rate in terms of percentage creep per hour the increase in the rate by doubling the minute initial rate does not appear to be alarming. However after a time such as 50,000 hours, the effect of doubling the hourly rate would be apparent in terms of total

elongation. The creep specimen must be kept in exact alignment during the long time of test under a load of dead weights applied through a system of levers. Manjoine (56) has reported a screw driven creep - rupture testing machine which obviates the necessity of an extensometer and lever arms, and weights are wholly eliminated. As was indicated in Figures 4, 5, and 6 a whole family of curves is needed for a particular material to see what behavior may be reasonably expected at different loads or different temperatures (54). With reference to curve A on Figure 2, it appears that there is no measurable elongation after the primary creep stage, but one can not conclude that the specimen will withstand this stress indefinitely. The initiation of the accelerating rate of creep ultimately leading to fracture cannot be accurately predicted for given conditions of stress and temperature from data based on tests carried out under higher stresses and temperatures (59).

In plotting creep, the curves on different steels frequently cross. Thus a material which has excessive creep at a lower temperature may be an improvement over the other material at a higher temperature (5). This is illustrated by the following data:

Stress for creep rate 1×10^{-6} in/in/hr.

Temperature	20 Cr-8.5 Ni	Other steel, Composition Not given
400°C.	44,000	24,500
500	29,000	22,000
600	14,000	15,000
700	5,500	4,000
800	1,500	2,000

(Other points noted in the above test:

1. extension less than .5% of gage length during first 24 hours.
2. during next 40 hours, no further extension
3. sensitivity of measurement, 10^{-4} inches on a 2 inch gage length).

STRESS, RUPTURE

The creep test was described as a test of materials for essentially long time application at constant temperature. As such the primary interest was in the secondary stage or uniform rate of creep. No particular interest was given to failure or rupture while discussing the creep test. In practice, however, the rupture life of a material is of very great interest to the designer of gas turbines. In the starting of a gas turbine, the temperature rises continuously until the steady operating temperature of the blades, discs, and other components is reached. Over a very short time, then, appreciable deformation may result from thermal causes as well as mechanical loading. Again, there is nothing to guarantee that the speed regulation device is going to remain operative at all times. Hence there may be periods of overload and higher stresses - stresses higher than those required for a minimum secondary creep (for example 1×10^{-5}% per hour). For this reason most highly stressed designs are based on creep and stress rupture data. Grant (60), working on the development of alloys suitable for long-time operation in gas turbines) set up the

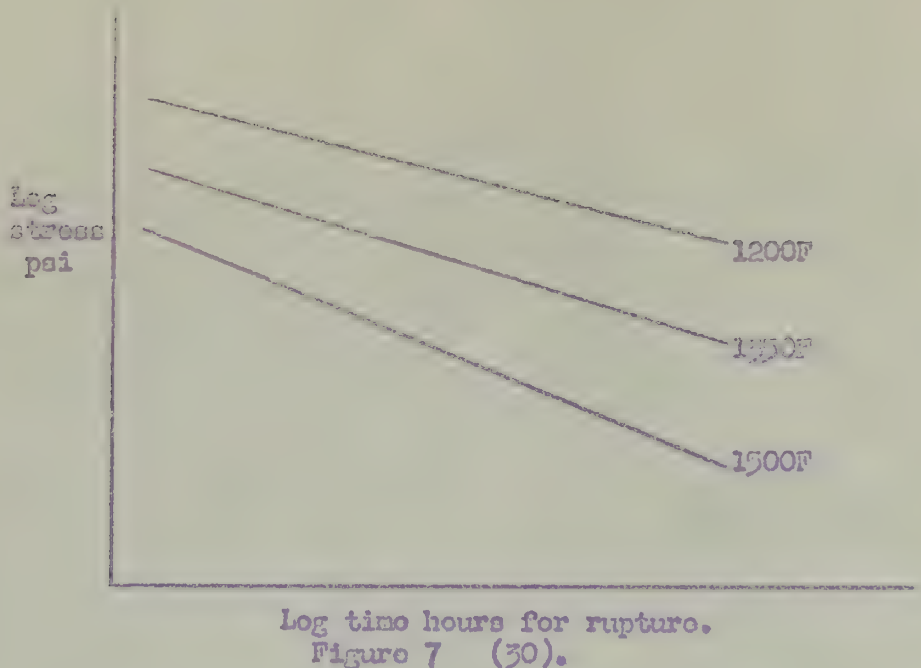


Figure 7 (30).

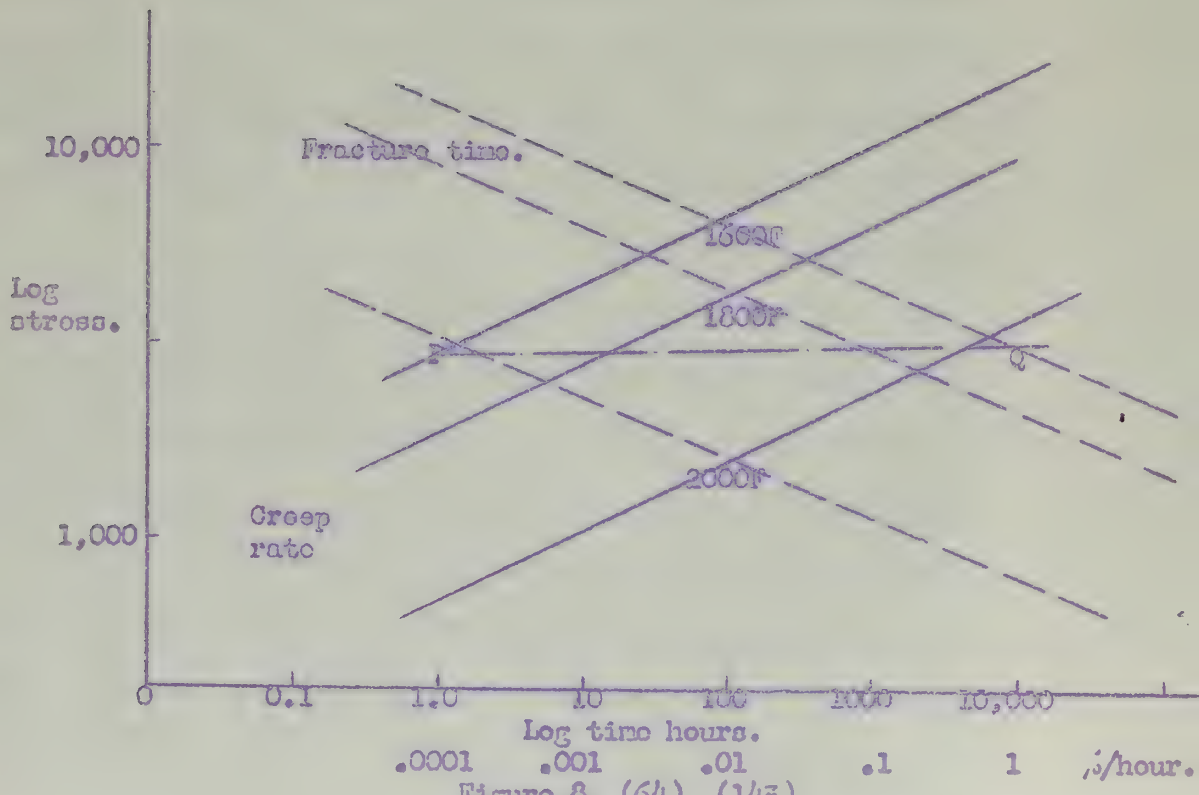


Figure 8. (64). (143)
Cast alloy 26 Cr 12 Ni .32 C .16 N

following standards as the aim of his program:

1. Minimum rupture time of 100 hours, 20,000 psi, 1500°F.
2. Creep rate $1 \times 10^{-5}\%$ per hour at 7,000 psi, 1500°F.

As a comparison of creep and stress rupture data, the following is given: (11) 1500°F.

Secondary creep 1% in 10,000 hours		Stress rupture strength 10,000 hours	Elongation at failure
S 816		14,500	9%
S 590	96,000 psi	11,000	11%
S 588		10-11,000	7%

Thus it must be noted above that while deformation under stress of all three alloys would be uniform under equal loading at a relatively low stress level, S 816 and S 590 would give greater protection against failure caused by overload or stress concentration not only due to their higher rupture strength, but also their higher ductility or ability to distribute or relieve stress concentrations by deformation.

The short time tensile test is really the first point of the stress rupture curve. Smaller loads are then chosen in succession to give longer rupture times. The tests are usually in the neighborhood of 200 to 2000 hours but the length of time appears to be a matter of personal choice. The variation of log time for rupture versus log stress is plotted in Figure 7 as a straight line at a given temperature. If the structure of the material changes or the surface becomes corroded the slope of the line will change. Thus there is great danger in extrapolating the straight lines to longer times and

lower stresses.

Epremian (22) and Grant (60) highly favor the stress rupture test as a preliminary test for choosing alloys for the testing of further properties. Gillett (27) however maintains that the designer or engineer may be steered away from materials useable at lower stress rates since the stress rupture test in effect gives information telling what will happen to the material under absolute abuse, loading to a degree which would not be allowable by correct design for long time service. Grant (60) in the development work previously mentioned, remarks that a good test for materials for high temperature alloys should embody the following four factors:

1. Ability to discuss favorable or unfavorable properties
2. Relatively rapid test time.
3. Reproducibility
4. Simplicity of apparatus and ease of interpretation of results.

According to Grant, the creep test if limited to stresses for about 2,000 hours which will not produce failure during that time requires costly and critical measurement equipment and then does not indicate the expected life or elongation at failure. In contrast to this, the stress rupture test seems to fulfill the four requirements above. The measurements made in this test are elongation versus time, elongation at failure, and rupture time. From the curve of time versus

elongation, a measure of the minimum creep rate in the secondary creep range is obtained.

Machlin and Nowick (49) have applied the theory of rate processes to stress rupture. Leaving out the thermodynamics and the mathematics, a brief resume is given of their work. The successful prediction of rupture time and the dependence on temperature for S 816 forged, S 816 cast, and low-carbon N 155 hot worked provided the reason for their recommending that rate-process stress-rupture equations should be used to interpolate and extrapolate stress rupture data for different temperatures. Thus the number of stress rupture tests necessary to get sufficient data for engineering design is reduced. The recommendation is made that a plot of log of rupture time versus stress be used instead of the log-log plots which have been used in the past as depicted in Figure 7. The semi log plot is necessary in order to obtain data to predict stress-rupture at additional temperatures by application of the rate process equation.

The above paper has received criticism on several points. Poncelet (discussion to 49) said that the stress-rupture data used covered only three full log cycles and this is too short a time to evaluate the use of semi log or log-log plots.

Evans (30) has indicated in the table below the ratios of creep strength to rupture strength for several alloys at 1200°F. and 1500°F. This data would seem to indicate that a factor of 50% may be

applied to the extrapolated rupture strength for 100,000 hour life to predict the creep strength for a secondary rate of 1% for 100,000 hours (1×10^{-7} inches per inch per hour). Another thing to note is that these ratios appear to increase as the temperature is increased.

	$\frac{C(1\% \text{ in } 100{,}000 \text{ hours})}{R(100{,}000 \text{ hours})}$		$\frac{C(1\% \text{ in } 10{,}000 \text{ hours})}{R(1000 \text{ hours})}$	
Type	1200°F	1500°F	1200°F	1500°F.
502	0.80		0.58	
446	0.50		0.376	
304	0.47	0.835	0.65	1.33
347	0.725		0.525	
316	0.56	1.0	0.512	2.2
310	1.14		0.755	1.0

(Rupture strength for 100,000 hours extrapolated)

One of the many annoying things about stress rupture data is the scatter in the results. Evans (30), reviewing data on fifteen of the older high temperature alloys, has remarked that the spread, may extend as low as 50% of the maximum value. Sweeney (21) thinks that most of the scatter in stress rupture data in the literature has been due to variation in aging prior to test, variations in chemical analysis, and variations in testing conditions. Accordingly, the Haynes - Stellite Company has undertaken a rather comprehensive program on groups of 64 stress - rupture bars cast from single heats. The purpose of this testing program is to arrive at a minimum value of stress-rupture so that the engineer can have positive data on which to base his design. A secondary objective is to determine

the cause for the scatter in the data. The quick failure of an occasional bucket is a rather serious matter since one fractured bucket may cause great damage to the remainder of the good buckets. A very distressing situation confronts the designer since after the one bucket which has failed prematurely out of 45 - 100 buckets has been replaced, the wheel will usually run the full expected life.

Figure 6 shows a spread purposely caused by heat treatment which could apply to large and small sections of the same material water quenched from the same hardening temperature. Avery and Wilks (48) made an attempt at a statistical analysis of stress rupture data at 1800°F. Only 18 cases were investigated and no conclusions were derived.

A common American convention as mentioned previously for reporting creep strength is the limiting creep stress for a designated

minimum (secondary) strain rate. Suppose that the limiting creep stress is 1% for 10,000 hours (0.0001% per hour). This is not necessarily a suitable design stress as the life expectancy above 1400°F at the limiting creep stress may be less than a year (approximately 8700 hours). Avery (48) has remarked that scatter bands for limiting creep stress and for 1000 hour rupture stress frequently overlap when plotted against temperature.

A very convenient way to plot creep and stress rupture data is to do the plotting on one graph as shown in Figure 8. Suppose it is desired to know the stress that will produce 0.1% creep in 1000 hours at 1600°F. The stress for these conditions would be given by point P, and then following horizontally across to the fracture time line for 1600°F to point Q and then down to the time scale, the time when the designated rate of creep will produce failure is obtained.

A recent paper by Grant and Bucklin (135) does much to clarify in the manner of rending a decision on some of the points previously discussed with regard to the stress-rupture test. In particular that above paper is concerned with extrapolation of stress rupture data. Grant and Bucklin made a great number of stress rupture tests on S 590 at test temperatures of $1200^\circ$to 1900°F and on S-816 at test temperatures of 1200° to 1500°F. These rupture tests had rupture times of 0.001 hours (3 seconds) to

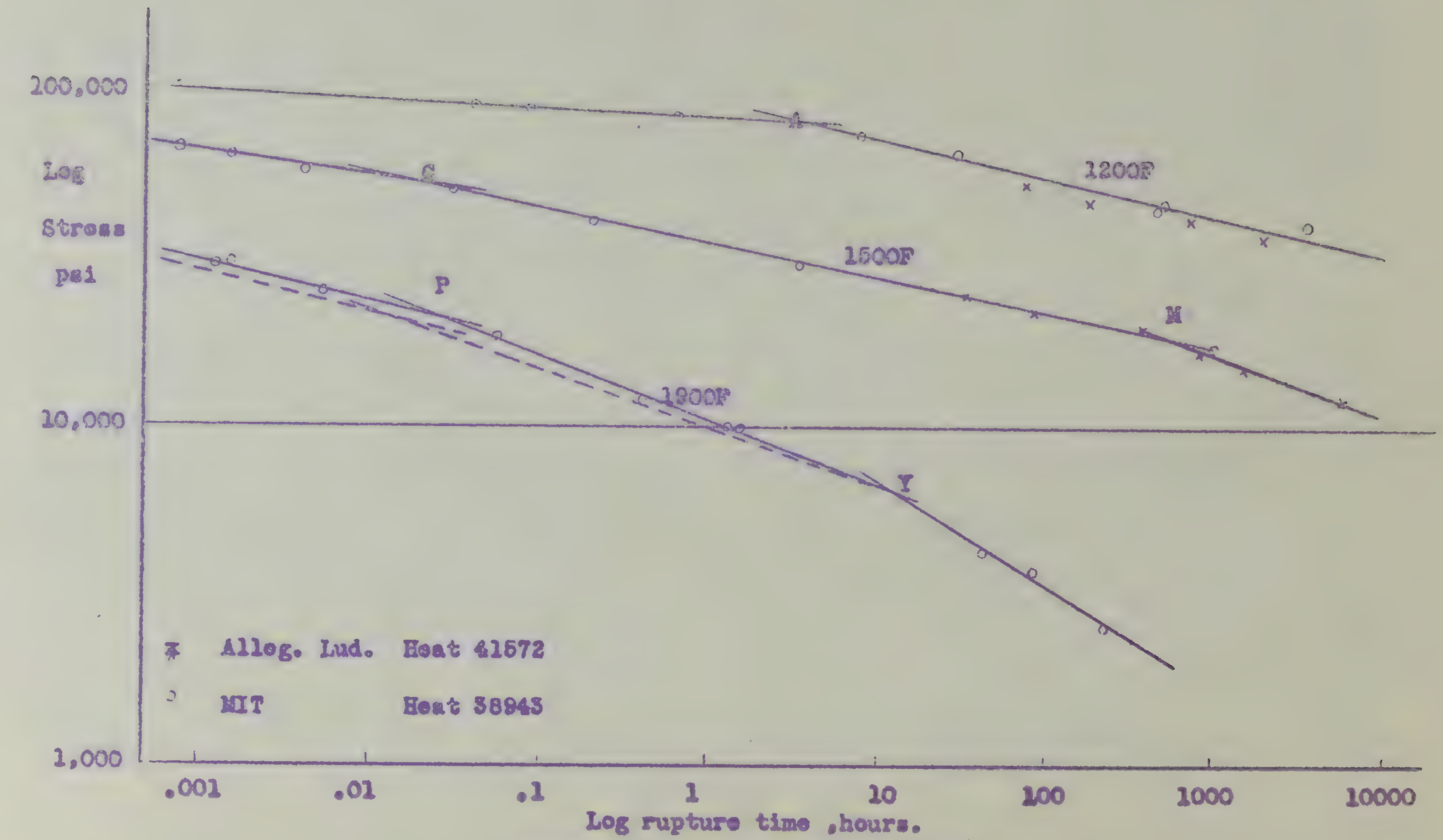


Figure 9. Alloy S-590. (135)

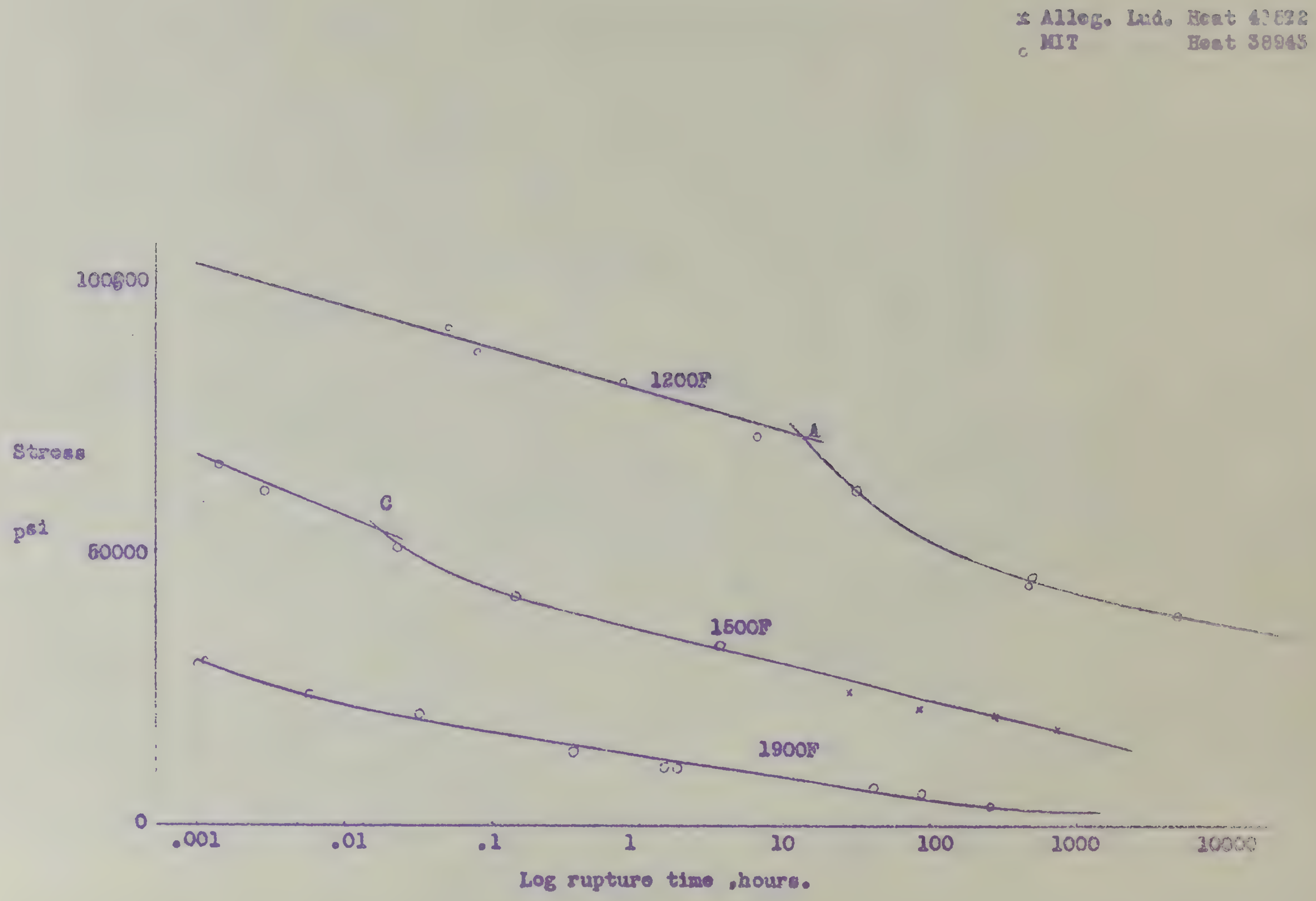


Figure 10. S-590. (135).

26,000 hours. Compositional and heat treatment variables were kept to a minimum. The heat treating schedules for these alloys were as follows:

S-590 Solution treat 2275°F for 1 hour, water quench
For all tests at 1200°F through 1400°F, age 16 to 24 hours at 1400°F.
For tests at 1500°F, age 16 to 24 hours at 1500°F.
For tests at 1600°F through 1900°F, age 16 to 24 hours at 1350° F.

S-816 Solution treat 2350°F for 1 hour, water quench
For all tests at 1200° through 1400°F, age 16 to 24 hours at 1400°F.
For all tests at 1500°F, age 16 to 24 hours at 1500°F.

Figure 9 shows the plot of log stress versus log rupture time for S-590 at temperatures 1200°, 1500°, and 1900°F. Figure 10 shows the plot of stress versus log time for S-590 at the same temperatures. It may be noted on the log - log plot Figure 9 that the data do not permit drawing unlimited straight lines. Grant emphasizes the point that the distinct breaks in the log-log plots should not be interpreted to indicate a physical or chemical change which is sharply discontinuous but rather a change of physical or chemical nature which has reached its maximum rate of change. These instability points depend on time, temperature, and strain rate and may be plotted to predict instabilities at other temperatures.

By means of three plots

1. Extrapolation of instability points log time versus temperature

2. Slope of straight line portions log slope degrees from horizontal versus temperature.

3. Plot of log stress for 10,000 hour rupture of extrapolated straight line curves versus temperature.

By using two of the three curves mentioned above it is possible to construct the predicted curve shown on Figure 9 (dotted line at 1900F). The unused curve above may be used as a check on the data figures obtained from the other two curves. The predicted curve shows fairly good agreement with the experimental curve.

In looking at the plot of stress versus log rupture life Figure 10, it is seen that beyond a certain point - time - stress - temperature - a continuous curve best fits the data. This information together with an attempt to extrapolate (temperature wise) curves according to the rate theory equation of Nowick and Machlin discussed earlier shows that rupture data should not be interpolated or extrapolated through the use of reaction rate equations. Furthermore, beyond a certain point in stress - time - temperature, stress versus log rupture time can not be plotted as a straight line.

In plots of "true creep elongation (first stage plus second stage) versus time Grant has shown that this "true elongation" decreases with increasing rupture time or decreasing strain rate, whereas the total elongation may or may not. This first stage plus second stage ductility is what is important to designers, since beyond the second stage of creep tolerances may be exceeded or failure may occur in a very short time.

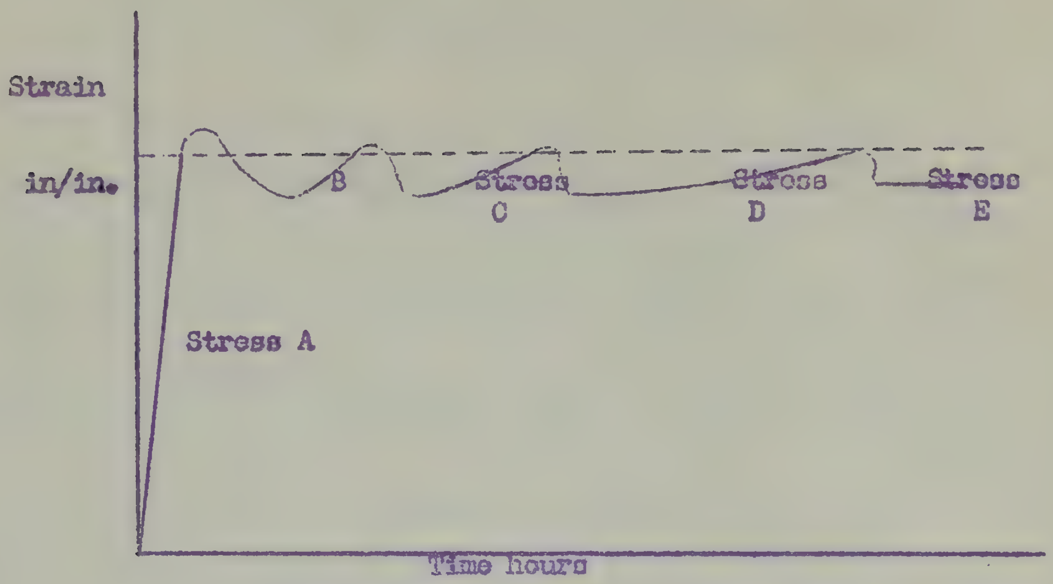


Figure 11 . Plot of typical creep relaxation . (30).

CREEP RELAXATION

Bolts for high temperature service are given the creep relaxation test. In this test the designer is interested in finding out how quickly and to what extent the initial elastic strain is converted to permanent plastic strain. The residual elastic strain is what keeps the bolt tight. If the residual elastic strain falls below the minimum design figure, the joint will leak. In conducting the test a plot of strain versus time is made in which a specimen of the alloy to be tested at a given temperature receives a series of loads which are decreased everytime a predetermined constant strain is reached. This strain corresponds to the initial bolting stress to which the stud will be "pulled up" at room temperature. Figure 11 illustrates the foregoing. In order to calculate the residual or relaxation stress which will remain in the bolt at temperature after any given time period, the logarithm of the secondary creep rate is plotted versus the logarithm of the stress. A straight line is obtained and then by means of a formula developed by Robinson, the relaxation stress may be calculated (30).

$$S = \sqrt[n-1]{\frac{bsn}{(n-1)\, r\, Et}}$$

S	residual stress in psi
n	slope of log stress - log creep rate plot
b	elastic ratio of system (10)
s	nominal stress to cause creep rate r.
r	nominal creep rate, 10^{-6} in/in/hr.
E	Young's modulus at temperature, obtained from slope of stress A
t	time for residual stress, hours.

SUMMARY

As was stated at the outset there is no established theory on creep and stress rupture. Actually there is not sufficient data of the proper kind to substantiate a theory. Most of the existent data is creep data or stress rupture data per se. The metals and alloys have not been tested with the idea of determining just what the creep test and stress rupture test is or does, but with the idea of finding out the limiting creep stress or rupture life of a particular metal or alloy at a particular temperature. Attempts to fit mathematical formulas to data obtained from these tests appear to be of not much utility principally on the point made above that there is insufficient data. The fact that stress-strain, time, and temperature must be considered simultaneously in discussing materials at high temperature is probably the big deterrent to the amassing of data. The great number of tests run by Grant and Bucklin (135) were on only two alloys S 590 and S 816. The particular findings on instability points in this paper refer to S 590 and S 816. Whether phenomena observed in "pure" low melting point metals is applicable to the multi-phased precipitation hardened high temperature alloys, remains to be seen as studies of structure are correlated with the creep and stress rupture test. Regardless of the state of the theory though, the creep test and stress rupture test are very valuable in determining suitability of alloys for high temperature service.

SUMMARY

As was stated at the outset there is no established theory on creep and stress rupture. Actually there is not sufficient data of the proper kind to substantiate a theory. Most of the existing data is creep data or stress rupture data per se. The metals and alloys have not been tested with the idea of determining just what the creep test and stress rupture test is or does, but with the idea of finding out the limiting creep stress or rupture life of a particular metal or alloy at a particular temperature. Attempts to fit mathematical formulas to data obtained from these tests appear to be of not much utility principally as the point made above shows that there is insufficient data. The fact that stress-strain, time, and temperature must be considered simultaneously in discussing materials at high temperature is probably the big detriment to the amassing of data. The great number of tests run by Grant and Bucklin (17) were on only two alloys S 590 and S 816. The particular findings on instability points in this paper refer to S 590 and S 816. Whether phenomenon observed in "pure" low melting point metals is applicable to the multi-phased precipitation hardened high temperature alloys, remains to be seen as studies of structure are correlated with the creep and stress rupture test. Regardless of the state of the theory though, the creep test and stress rupture test are very valuable in determining suitability of alloys for high temperature service.

CHAPTER III

STRUCTURE

Variations in structure of a metal or alloy cause variation in physical properties. Changes in crystal structure, solution and coalescence of precipitates at elevated temperatures, precipitation of minute particles at lower temperatures, typical cast or wrought types of structure, graphitization of carbon grain orientation, and grain size are some of the factors which may be discussed under the broad term "structure." X-Ray diffraction and metallographic examination are the principal means of studying structural variations. There is much written in the literature regarding the relative merits of cast and wrought materials for gas turbine blades. Henry (7) has remarked that the designers decision as to whether a wrought or a cast part will be used is influenced by a number of factors in addition to the primary comparison of physical properties. Such factors as production speed, cost, number of rejections, reproducibility, and inherent ease or difficulty of processing must be considered. These last named factors are economic in nature and are not discussed in this report. In fact the primary reason for discussing structure at all is to show the effect of structure with regard to the various mechanical tests.

Cross and Simmons (47) working with 46 different alloys of Cr - Ni - Fe, Cr - Ni - Co - Fe, Co - Cr, and Co - Cr - Ni families arrived at the following general conclusions as to cast and wrought

materials:

1. The cast alloys of the Co-Cr and Co - Cr - Ni base type show better stress rupture strength than the wrought alloys at the higher temperatures.

2. The better forged alloys as a group had better creep resistance than the cast alloys near 1350°F.

3. Cast alloys showed lower impact resistance at both room temperature and at 1500°F than wrought alloys.

4. Cast cobalt - base alloys show consistently better stress rupture properties, than wrought cobalt-base alloys. Although cast cobalt - base alloys are referred to as relatively brittle alloys at room temperature, turbine buckets of this material have been known to bend rather than snap off when rubbing contact occurred during service test at operating temperature.

The above statements probably are subject to the particular influences of melting, casting, working, and heat treating variables. Thus these statements can not be used as hard and fast rules, but do supply some useful information. In other words for a plant operating under a fixed set of processing variables, mechanical tests would have to be conducted on the respective forged and cast materials to be positive about which has the better properties.

For example, there were several aging and heat treating treatments applied to the alloys above (47). The wrought alloys showed best stress rupture and creep values when tested in the solution - treated and aged condition rather than in the as - forged or as - rolled and aged condition. Heating from 30 minutes to 4 hours at 2150 to 2300 F effected solution of precipitated phases. Quenching prevented precipitation of these carbides or intermetallic compounds.

Aging at 1350 to 1600F produced precipitation in a fine dispersion which on testing indicated increase in high-temperature strength. The high temperature solution treatment and aging for those alloys not undergoing phase transformations resulted in considerably coarser grain size, increased strength, and lowered ductility. Hardness measurements after various aging times indicated that after 50 hours of aging there was little change in hardness. Accordingly, all test specimens were aged 50 hours at the intended test temperature in order to have alloy structure and properties that would exist after the alloy had been in service for a time. These 46 different alloys of the Cr - Ni - Fe, Cr - Ni - Co - Fe, Co - Cr, and Co - Cr - Ni families (47) all age in the temperature range 1350 - 1600F. Sometimes the aging effect is very marked. For example vitallium and alloy 61 after being aged at 1350°F prior to testing had yield strengths at 1000 and 1200F almost double those in the as cast state. Gordon (52), Scott and Gordon (110), Dobkin (54), Foley (67), Grant (75) (76) (60), and Henry (12) are a few of the investigators who have written papers in which aging and precipitation effects are prominently discussed. With reference to iron base materials Dobkin (54) has said that the carbide forming elements improve elevated temperature strength by formation of a second precipitating compound after high temperature has caused iron carbides to coalesce to a degree in which they are no longer very effective as a hardener.

These second precipitating carbides agglomerate slowly even at temperatures up to 1800°F and thus remain small, numerous, and very effective as a hardening dispersion at high temperatures. Dobkin states that further proof of the fact that it is the carbides of these elements which produce the resistance to softening is, that without the presence of sufficient carbon, a considerably greater proportion of dissolved high temperature strengthening elements is necessary to produce similar creep resistance.

Grant (76) has noted, in alloy 28 V-0 that after a solution treatment at 2300F had dissolved all of the massive carbides, the aging constituent was very uniformly distributed throughout the grain, and there is a non continuous agglomeration of a phase (unidentified) at the grain boundary. Air cooling from 2300 F causes an extremely heavy grain boundary precipitate and less ductility than above by furnace cooling. The fact that a furnace cool gave the optimum structure in alloy 28V-0 is at odds with the general method of heat treatment in age hardening. Binder (20), for example, has noted that the rate of cooling after solution heat treatment controls the distribution of the precipitate with respect to grain areas and grain boundaries. Accordingly, the rate of cooling should be as rapid as possible to guard against excessive precipitate forming at the grain boundaries. Also the temperature of formation of the precipitate influences the size of the particles of precipitate due to relative rates of nucleation and growth. This case, then, of

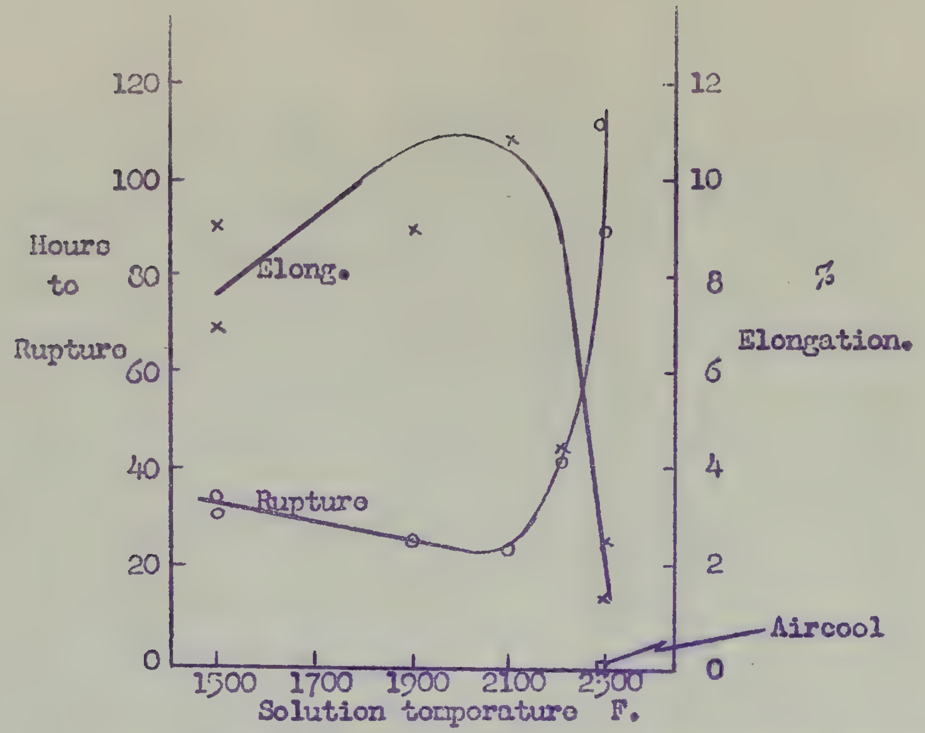


Figure 12. Effect of solution treatments on rupture properties of alloy 28V-0 , $\frac{1}{2}$ hour at temperature followed by furnace cooling.No aging. (20,000 psi,1500F) (76).

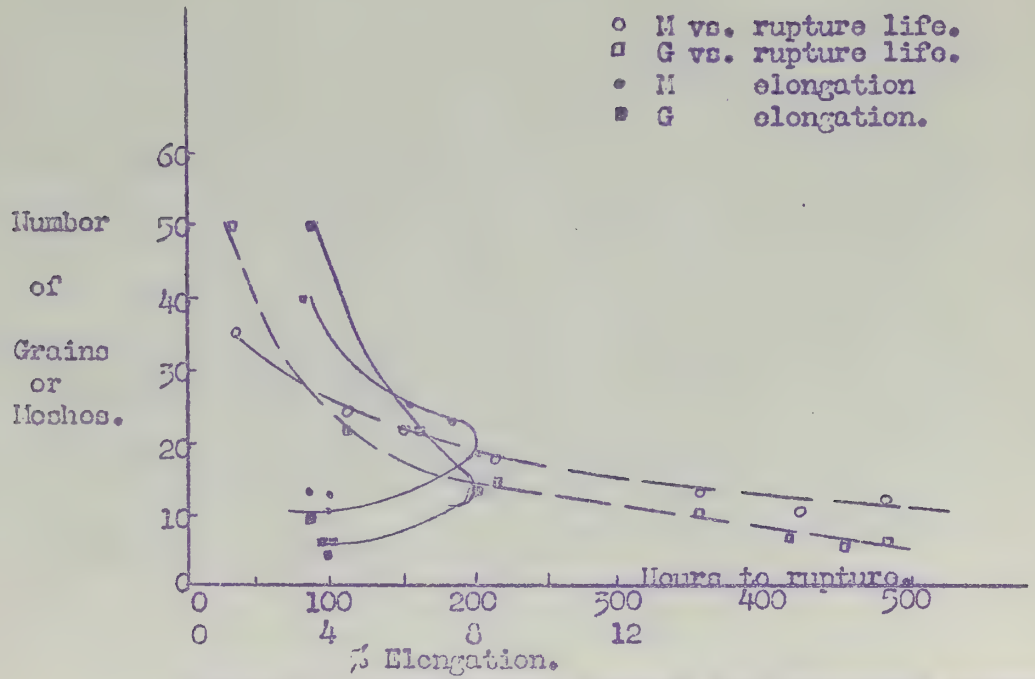


Figure 13. Relationship between the structural elements of alloys 100VT 2-2 and rupture properties at 30,000 psi and 1500F. (77).

28 V-O and several of the other vitalliums is one of the many anomalies appearing in the high temperature alloys which make it rather difficult to make general statements of universal application. Figure 12 (76) illustrates the decrease in ductility accompanying an increase in rupture strength as a function of solution temperature. At 2200 F it was noted by microscopic studies that the carbides had been partially taken into solution and the aging constituent is distributed more uniformly throughout the matrix.

Aging for 48 hours at 1350 F decreases the ductility (% elongation) of the vitallium - type alloys as compared to as cast properties. (60) It would seem that the hot or cold tensile test does not indicate a very important measure of ductility since the important factor is the rate of decrease of ductility with time. With the stress rupture test, the aging effects of decreased ductility is shown by making % elongation measurements during the progress of the test. Grant and Lane (75) working on four alloys of varying compositions which are given below for comparison found that all four alloys, as cast, contain at least 3 of what appear to be 4 distinct phases:

1. The continuous phase (a face - centered cubic and or hexagonal structure, called austenite).

2. A globular or dendritic material referred to as a carbide.

3-4. Two remaining constituents, present in much smaller

amounts in part resembling the above carbide.

	Alloys									
	C	Mn	Si	Ni	Cr	Co	Mo	W	Ta	Fe
25 V-0	0.25	1.0			23	68	6			
45 R-6	.45	1.0		32	26	33	6			
100NT-2	1.0	1.5	0.5	30	20	20	3	2.2	2	Bal.
111VT2-2	1.1				23	67	6		2	

In the two low carbon alloys above, massive carbides constitute only a small area of any cross section and do not form a complete path through the structure. Here the aging benefits imparted to the so called "austenitic" matrix as shown by stress rupture and creep tests were large. In the above alloys, the first definite precipitate noticed occurred at 1350F. Maximum rate of aging in the two low carbon alloys was at 1550 - 1600 F and in the two high carbon alloys at 1600 - 1700 F. Agglomeration of the precipitate began at 1700 F. At 2100 - 2200 F almost all of the precipitate was back in solution. The carbides were observed to have a melting point slightly lower than the rest of the alloy. The surprising result of the above investigation is the discovery that alloys differing so much in composition (C .25 - 1.1%, Co 20 - 68%) behave so much alike. Iron occurred in impurity amounts to 20%. Yet the temperatures for an aging precipitate to form and go back in solution were very close.

Strain hardening has an accelerating effect on precipitation of phases from solid solution. Thus if an alloy is in a supersaturated condition, strain hardening brings about precipitation at temperatures

where the influence of the precipitate on mechanical properties would normally be ineffective because of the slow rate of precipitation.

With certain high chromium steels, a hard, brittle, and non-magnetic constituent called the sigma phase of the general formula FeCr, but having a broad composition range, tends to form in the ferrite of these steels when the steels are heated for a time in the temperature range 925 - 1750F. (67) (12) (146). The formation of this constituent in appreciable amounts causes an alloy steel to lose its ductility to an extent that may make the steel unsuitable for a particular application. Other manifestations of the sigma phase are increase in hardening and decrease in notch resistance.

Instability points are the names given to the breaks in the straight line relationship of the plot of log stress versus log time in Figure 9. Grant (135) remarks that these instabilities may result from one or more of the following processes: (135)

1. "Change from typical low temperature type deformation to high temperature type deformation. This is evidenced by the change from slip flow, strain hardening, and transcrystalline failure to essentially slipless flow, annealing, and intercrystalline fracture
2. Oxidation and corrosion
3. Overaging, agglomeration, and resolution of a precipitate.
4. Other structural changes such as recrystallization, with or without grain growth.
5. Phase changes such as the formation of the sigma phase."

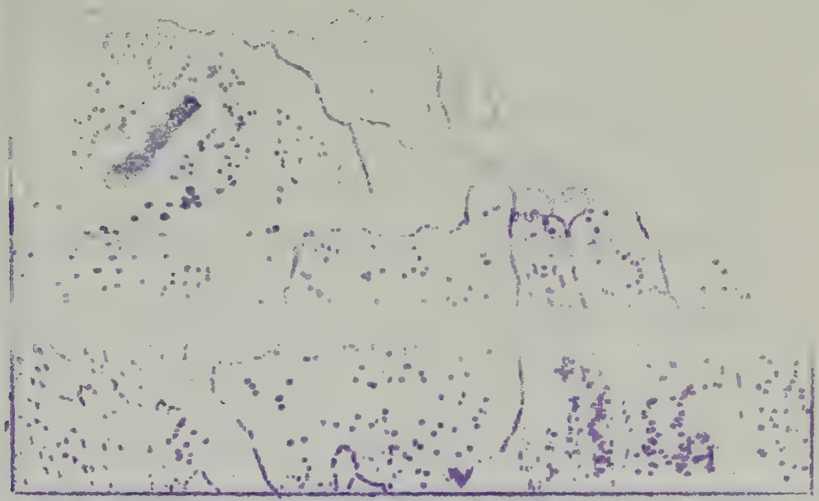


Figure 14. S-590, ruptured after 0.016 hours at 75,000psi and 1350F. Transcrystalline failure. Electrolytic ammonium hydroxide etch. 100X. (135)

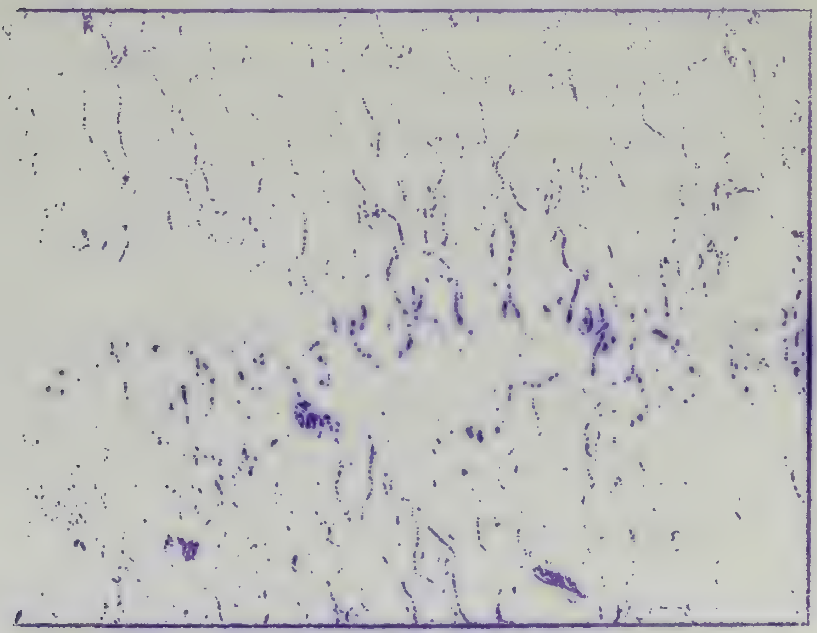


Figure 15. S-590, ruptured after 29.9 hours at 60,000psi and 1200F. Intercrystalline fracture. Electrolytic ammonium hydroxide etch. 250X. (135)

By microscopically examining fractured specimens, Grant (135) found that at rupture times immediately to the right of AC a failure which begins at a grain boundary diminishes the cross section of the test bar, increasing the stress and strain rate, and thereby causing the remainder of the bar to fail in a transcrystalline manner. According to Grant this might make it appear that the failure was transcrystalline, but the presence of frequent grain boundary cracks at the surface of the specimen in these instances indicates the transition to high temperature type (grain boundary) of deformation and failure. Figure 14 and Figure 15 illustrate transcrystalline failure at a rupture time to the left of AC and intercrystalline failure at a rupture time to the right of AC on Figure 9. The second series of breaks in the curves, of which only two breaks, M and F, are shown, is attributed to the start of overaging. As in the case of the breaks at AC, the breaks MF are shifted to shorter times with increasing temperature. (F at 1900°F to the left of M at 1500°F.)

The third set of breaks indicated by point Y at 1900 F, from an examination of the microstructure of failed bars, is probably due to several factors. Oxidation is showing its effect. Decarburization is noted at intercrystalline cracks.

Badger and Sweeney (45) have suggested the development of an alloy which would not exhibit some of the problems due to aging that have been described above.

There are some structure in carbon and low alloy steels that have not been fully identified (68). It is small wonder then that the precipitated compounds in the high temperature metals have not been fully determined. Metallographic examination of hot metal surfaces is proceeding as a result of the building of special microscopes. Instead of the essential optical components of the objective being lenses, these components are mirrors (74). The real image of the subject is projected on the stage of the microscope (70). In this way, the optical system may be protected from radiation from the material being examined by moving the optical system to a greater distance from the subject. Long heating runs can be condensed into a short period of time of viewing by taking motion pictures and then running the film through the projector at varying speeds (70).

X-Ray methods have been used and are being used to study structure of the high temperature alloys. The crystal structures of 10 wrought heat resisting alloys at elevated temperatures have been compared with their crystal structure at room temperature by J. H. Kittel (69). The alloys studied by means of a Geiger - counter X-Ray spectrometer modified for high temperature use were S-816, S-590, Hastelloy B, 19-9 W-Mo, N-155, 16-25-6, K-42-B, Inconel X, Nimonic 80, and type 347 stainless steel. All of the above alloys for example, except 19-9 W-Mo retained their room temperature crystal structure up to a temperature of 1800 F. 19-9 W-Mo transformed from

a two phase mixture of B. C. C. and F. C. C. to a single phase of F. C. C. between the temperatures of 1400 - 1450F. Thermal dilation studies were made up to 2150 F on these alloys. None of the alloys showed discontinuities ascribable to a phase change.

By means of the X-Ray diffraction apparatus in conjunction with a Geiger - Muller counter indicated above, it is possible to detect phase changes taking place in a time of one second (66).

Recently an X-Ray diffraction investigation was made of the minor phases of 20 high temperature alloys (65). The identification of these minor phases in an alloy is important in determining the distribution of the elements of the alloy among the phases since knowledge of the manner in which a specific element is distributed among the various phases of an alloy will aid in the evaluation of the function of that element in the alloy. This information would also help in selecting the percentage of an element required in a new alloy containing the same types, but not the same quantities of minor phases.

In this work, the minor phases were concentrated by electrolytically digesting the alloy in a bath of a reagent that selectively dissolves the matrix. The residue deposited on the bottom of the bath is highly concentrated in minor phases. If several specimens of an alloy are subjected to electrolytic attack by different reagents, each reagent will dissolve the alloy phases in varying amounts or proportions. Since these phases present were unknown, 12 reagents

were used in an effort to concentrate each of the minor phases as a residue in at least one reagent. By a comparison of the patterns of the several residues obtained, it was possible to identify the phase by the presence of its interference lines. In this manner the following minor phases were determined: CbC, CbN, TiC, TiN, Cr_7, C_3, $Mo_{23}C_6$, and Mo_6C. All residues were not identified. Much work remains to be done in this field.

Usually the ductility (% elongation) of a fine grained material is superior to that of the coarse grained material regardless of time of testing. (11). With regard to creep strength and stress rupture strength though, the coarse grained material usually shows higher values than the finer grained. Clark and Freeman (78) found that coarse grained 18-12 Cb, 25 Cr - 20 Ni, and 25 Cr - 12 Ni were much stronger than their fine grained counterparts. Grant (80) has said, as a result of the examination of countless failures and near failures, that ruptures can originate internally or at the surface. Initially, it is claimed, all failures start at the grain boundary above a certain temperature. Thus the less grain boundary material there is, the less is the probability of the initiation of failure.

Grant (77) made a study of the relationship between the rupture properties and the structural elements in alloys of the Co - Cr - Mo - Ta system. The mesh number and the grain size of the were used in an effort to determine some of the minor phases in a residue in at least one example. By a comparison of the patterns of the several residues obtained, it was possible to identify the phase by the presence of its characteristic lines. In this manner the following minor phases were determined: TiC, TiN, ZrC, NbC, Cr_7C_3, $Cr_{23}C_6$, Fe_3Mo_3C, and Mo_2C. All residues were not identified. Much work remains to be done in this field.

Usually the ductility (% elongation) of a fine grained material is superior to that of the coarse grained material regardless of time of testing. (113). With regard to creep strength and stress rupture strength though, the coarse grained material usually shows higher values than the finer grained. Clark and Freeman (73) found that coarse grained 18-8 Cb, 18 Cr - 25 Ni, and 25 Cr - 12 Ni were much stronger than their fine grained counterparts. Grant (81) has said, as a result of the examination of crackless failures and near failures, that ruptures can originate internally or at the surface. Initially, it is claimed, all failures start at the grain boundary above a certain temperature. Thus the less grain boundary material there is, the less is the probability of the initiation of failure.

Grant (77) made a study of the relationship between the rupture properties and the structural elements in alloys of the Co - Cr - Ni - Fe system. The main phases and the grain size of the various specimens were determined. Mesh number is taken as the number of carbide intersections at a magnification of 400X with a gage length in a micrometer eyepiece of 0.04 inch. The number of carbide intersections with the gage length is counted in several different directions. Wide differences in mesh number can be obtained depending on grain orientation and the manner of cutting through a particular dendrite. Figure 13 illustrates the relationship between grain size and mesh number and the rupture properties of alloy 100 VT 2-2 at 30,000 psi and 1500 F. If the maximum hot ductility (8%) is desired for a particular application, it is noted that 200 hours rupture life may be attained with this ductility at 30,000 psi and at a temperature of 1500 F. To get these properties, then, in a ¼" diameter test section (for which the graph was plotted) it is necessary to have 19 carbide intersections per unit of gage length with about 16 grains in the entire section. There are a number of things to be careful of when evaluating the above work. Grant mentions that the work involved only small sections where the mass of the metal is small in comparison to the mass of the refractory material of the mold. There are no zones of columnar crystals which give way to equiaxed crystals at the center of the specimen as might be noted in large sections. Round cylindrical specimens and molds were used. Sweeney (21) amongst others has remarked that grain size and grain orientations in a cast turbine blade are much different than in a round specimen. Cross and Simmons (47) tested diamond-shaped

specimens with sharp edges and thin sections approximating those in a turbine bucket. The diamond shaped specimens showed rupture properties in the lower portion of the range shown by the round test specimens.

Grant further remarks that the materials used were 2 or 3 phase alloys. What the effect of single phase material would be is not known.

Although a coarse grained material is desirable for great strength at high temperatures due to having least grain boundary area, grain orientation becomes a very important factor. Grant (42) by raising the temperature of the molten metal and the temperature of the mold several hundred degrees cast a 1% C vitallium specimen in which any cross section of the .250 inch diameter specimen was a single grain. A particular specimen gave a rupture life of 1500 hours at 30,000 psi and 1500 F. Other specimens cast under the same conditions

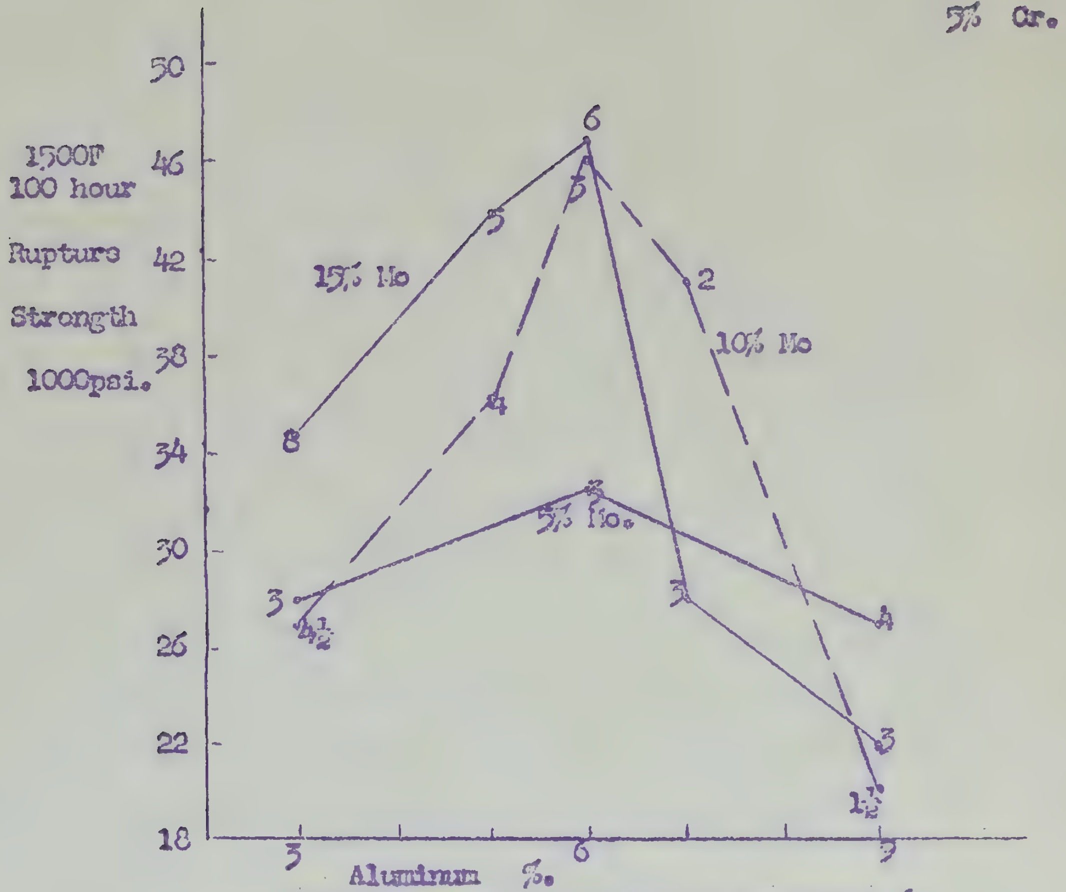
showed rupture times of 180 - 300 hours. The grain orientation of the 1500 hour specimen was ideal whereas the other bars had grain orientations which permitted early failure along the grain boundary at angles approaching 45° due to shear stresses. Since the control of grain orientation is strictly a function of probability, Grant (42) (77) recommends 12 to 16 grains per ¼" cross section for 1% C vitallium alloys in order to get some reproducibility in stress rupture properties. In the case of 12 or 16 grains per ¼" cross - section only 1/12 or 1/16 of the total strength or weakness is due to a perfectly oriented or poorly oriented grain.

It is difficult to evaluate the effect of a particular alloying element per se on mechanical properties of an alloy unless the structure of the material remains essentially the same before and after the incorporation of the particular metal into the alloy.

elapsed rupture times of 150 - 300 hours. The grain orientation of the 1500 hour specimen was ideal whereas the other bars had grain orientations which permitted early failure along the grain boundary at angles approaching 45° due to shear stresses. Since the control of grain orientation is strictly a function of probability, Grant (42) (77) recommends 12 to 15 grains per 1" cross section for 18 8 vitallium alloys in order to get some reproducibility in stress rupture properties. In the case of 12 or 15 grains per 1" cross - section only 1/12 or 1/15 of the total strength or weakness is due to a perfectly oriented or poorly oriented grain.

It is difficult to evaluate the effect of a particular alloying element per se on mechanical properties of an alloy unless the structure of the material remains essentially the same before and after the incorporation of the particular metal into the alloy.

This is to say that perhaps a phase change or precipitation came about as a result of adding the metal. Therefore the question may be added, "Is the increase or decrease in the mechanical property due to the phase change or precipitation or to the inherent alloying effect of the metal addition?" This question may readily be answered for steels tested at room temperature by using another alloying element which accomplishes the same change in structure. Usually, it will be found that the structure change is the predominating influence on the mechanical properties although the order of magnitude of the change will be related to the particular alloy addition, and the amount of the alloy addition needed to effect the structure change will depend on the particular element. Nevertheless, there are several general statements which may be made regarding the effect of certain elements on stress - rupture properties, for example. Grant (42) (34) with a close control of the structure of the alloy has noted the effect of carbon in about 8 different alloy systems and the effect of tantalum and nickel up to 8% in vitallium alloys. Gross and Simmons (47) have noted that the addition of about 20% cobalt to Ni - Cr - Fe alloys increases the high temperature strength markedly over Ni - Cr - Fe alloys without cobalt. Grant (42) also noted that certain cobalt base alloys gave higher high temperature strength than certain chromium base alloys. In general, most investigators agree that cobalt increases the high temperature strength of an alloy.



Note: Figures at each point represent % elongation in 2 inches.

Figure 16. Effect of aluminum content on the 100 hour rupture strength of Ni base alloys containing 5% Cr and 5,10,15 % Mo. (31).

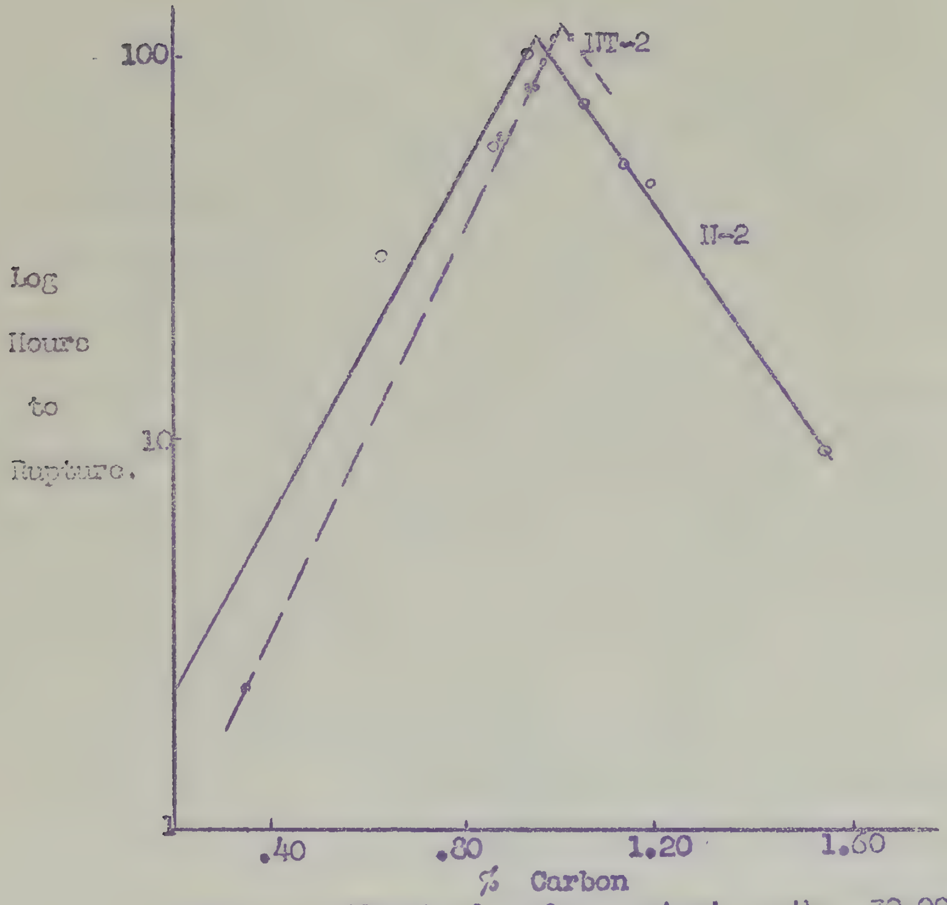


Figure 17. Effect of carbon content on the 30,000 psi rupture life of Ni-Cr-Co-Fe base alloys at 1500F. (60).

Guy (31) has conducted a survey of 100 hour and 1000 hour rupture strengths at 1500 F on nickel base alloys in which aluminum, molybdenum, and chromium were varied. This paper has a number of plots illustrating the effect of alloying elements on the stress rupture strength. Figure 16 is one of these plots showing the effect of aluminum content on the 100 hour rupture strength of alloys containing 5% chromium and 5, 10, or 15% molybdenum.

In Figure 13 was indicated the relationship between mesh number and rupture strength and mesh number and elongation. Grant (77) found that the carbon content (.40 - 1.20%C) does not control the carbide mesh spacing in Co - Cr - Mo - Ta alloys, but controls only the amount and massiveness of the carbide. With Ni - Cr - Co - Fe base alloys, Grant (60) found a straight line relationship between the log of the rupture life and percent of carbon. In Figure 17 (60) is shown the effect of carbon content on the 30,000 psi rupture life at 1500 F. The increase in rupture strength with increase in carbon, the peak in the rupture strength, and the decrease in rupture strength after the peak may be explained on the basis of changing structure as shown by metallographic studies. As the carbon increases, the ratio of volume of carbide mesh to matrix increases. Finally, the carbide mesh assumes a large enough proportion of the volume of the alloy to carry the bulk of the load. At a certain carbon content the carbide mesh structure becomes too massive and or continuous.

Beyond this point the rupture strength begins to fall off.

The very sketchy treatment given to structure above is by no means complete. The primary purpose of discussing structure was to show the effect of structure on high temperature properties, particularly stress rupture. Much work remains to be done in the manner of some of the papers mentioned here. At the present time, there is not even sufficient experimental evidence to make good qualitative or empirical observations or decisions.

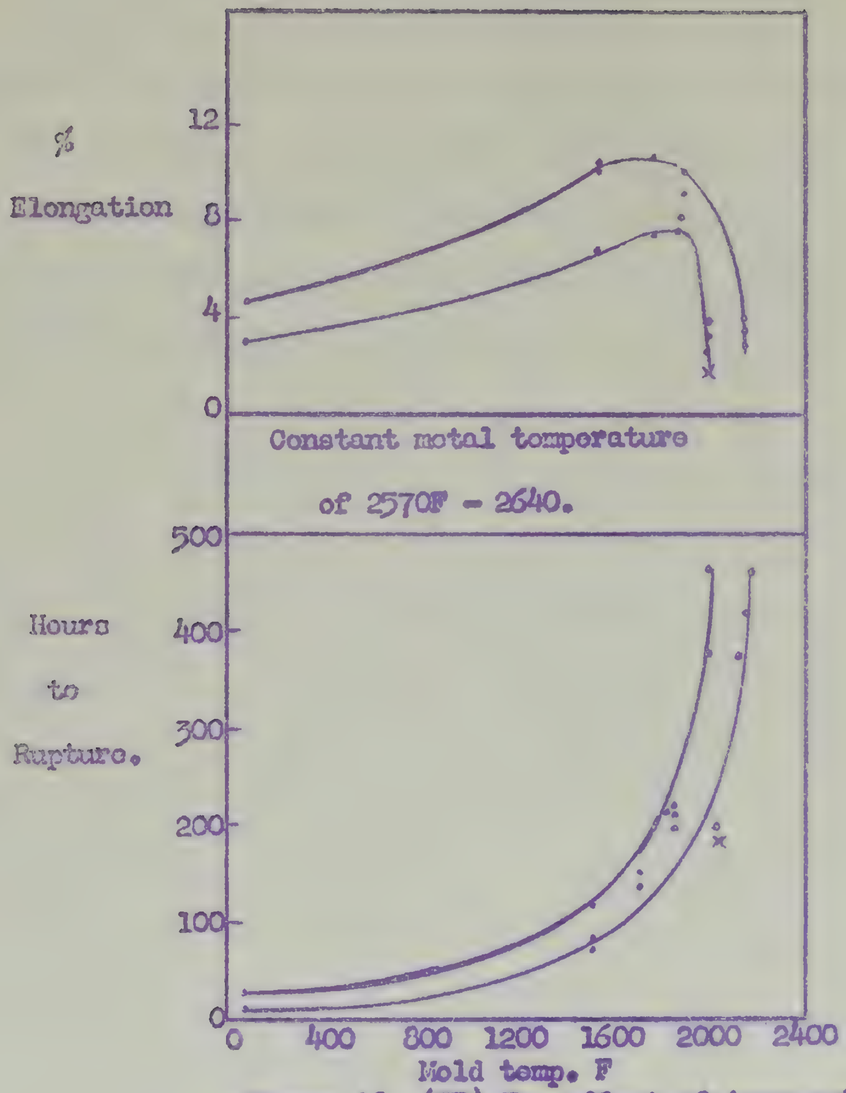


Figure 18. (77) The effect of increasing mold preheat temperature on the 30,000psi and 1500F rupture life and elongation in alloys 100VT2-2. Crosses show single grain fracture.

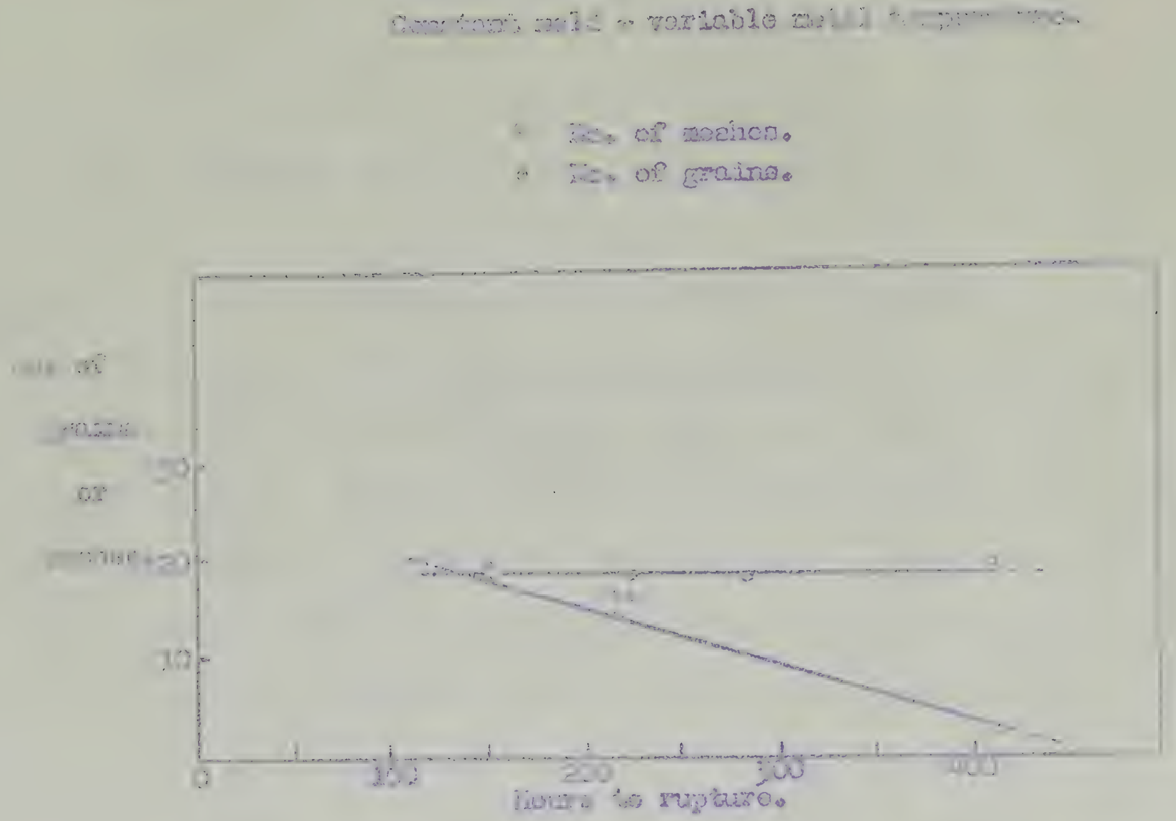


Figure 19. (77) The relationship between the structural elements of alloys 100VT2-2 (as determined by variation in the metal pouring temperature) and the rupture properties at 30,000 psi and 1500F.

CHAPTER IV

MELTING, PROCESSING AND FABRICATION

In the section on structure some mention has been made about the influence of melting and casting variables on the microstructure and consequently on the physical properties of some of the vitallium alloys. A few additional remarks are added here.

Grant (77) has reported that occasional cast high temperature, high strength alloys of the Co - Cr - Mo - Ta type failed to show reproducible rupture properties by yielding either excessively high or low test data. Accordingly a test program was initiated to study the effect of mold preheat and metal casting temperature on the rupture properties. Figure 18 shows the effect of increasing the mold preheat temperature on the 30,000 psi - 1500 F rupture life and elongation in alloy 100 VT 2-2. Note the wide spread of rupture life values with mold preheat temperatures greater than 1850 F.

Grant (76) found that the number of grains per section decreases with increasing metal temperature while the carbide spacing remains constant. The above is reflected in Figure 19 which summarizes the relationship between the structural elements of 100 VT 2-2 (as determined by variations in metal - pouring temperature) and the rupture properties at 30,000 psi and 1500 F.

Much work remains to be done in the investigation of the influence of processing variables on physical properties. The Statistical Research Group at Columbia University under an NDAC contract, could find no correlation in vitallium among composition, heat treatment, rupture life, and ductility on the basis of data obtained from various sources. This failure has been attributed to variations in mold and metal temperatures.

At Allegheny Ludlum Steel Corporation, Brackenridge, Pennsylvania (104) melts of S-588, S-590, S-816 are made in a 2½ ton or 10 ton basic electric arc furnace. The heats are finished under a basic reducing slag and are poured at 2700 - 2900 F into big end up molds 6 - 16 inches square weighing from 400 - 3200 pounds. In order to secure sound ingots a non consumable electrode arc has been successfully used to keep the top of the ingots hot. These data are included in order to make a comparison as to size in commercial applications with the precision investment casting (sometimes .25" cross section) used by Grant in the experiments outlined above.

Heat Treatment, Working, and Fabrication

Disc material must be forgeable and buckets, if wrought buckets are to be used, must be amenable to working or machining. All of the high temperature alloys are difficult to hot work and to hot - cold work in the range 1200 - 1700 F because of their high temperature strength. This fact is one of the principal problems in the development of high temperature wrought materials. Freeman, Reynolds, and White (46)

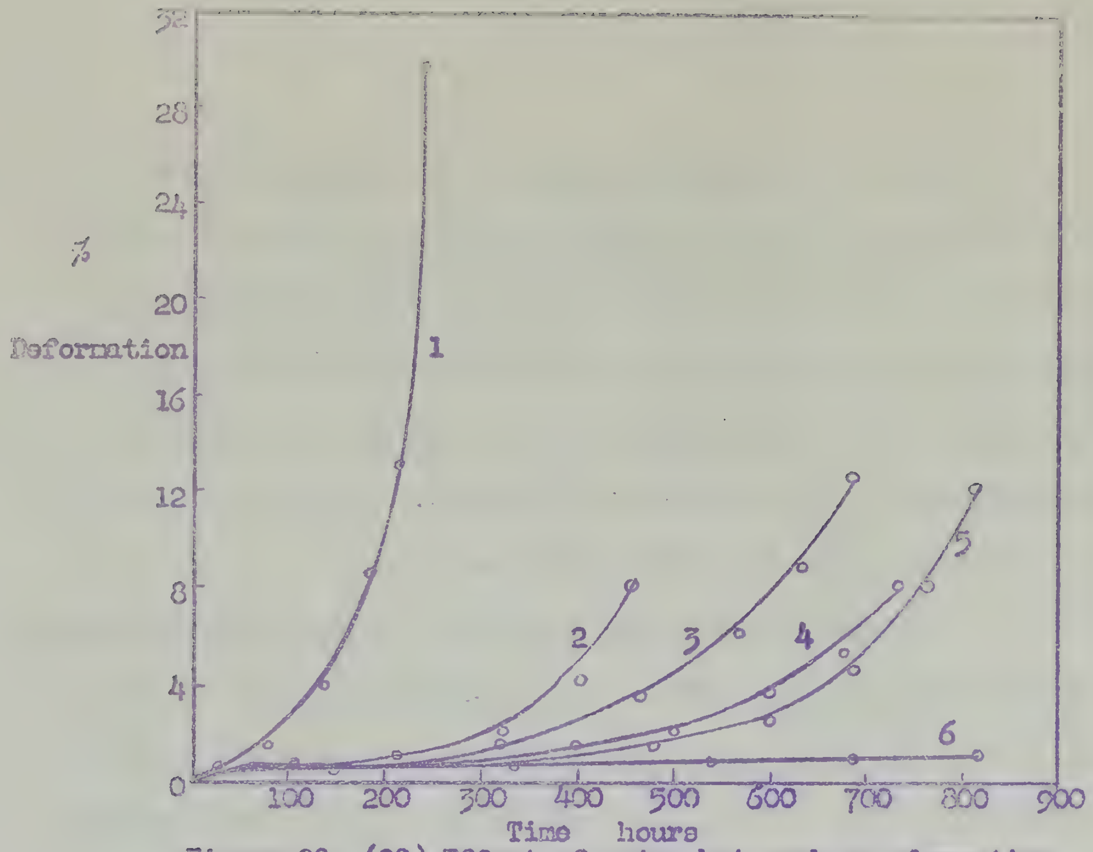


Figure 20. (20) Effect of prior hot work on elongation of N-155 alloy at 1500F under constant stress of 8000 psi.

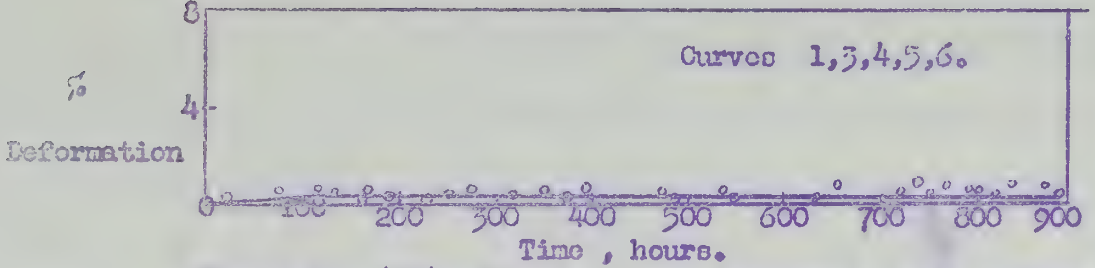


Figure 21. (20) Effect of heat treatment on elongation of N-155 alloy at 1500F under constant stress of 8000 psi.

found that it was difficult to reproduce commercially the degree of hot - cold work on experimental bar stock. The effect of a given heat treatment on light sections may be radically different than on a heavy section. Much of the data accumulated to date with regard to working has been obtained from controlled experimental conditions on small bar stock.

Heat treatment has been mentioned earlier with reference to solution treatment and to stabilizing heat treatments at temperatures above that of service temperature. With some materials, it has been noted that better creep and stress rupture properties are obtained when the material is stabilized at say 1350 F for eventual use at 1500 - 1600 F.

N-155 is probably one of the most widely used forgeable materials for the wheels of gas turbines. The effect of hot - work on the elongation of N-155 alloy at 1500 F under a constant stress of 8000 psi is shown in Figure 20. The resistance to creep may vary widely depending upon conditions during hot working. Yet, by glancing at the table below, it may be noted that none of these conditions were abnormally different (20).

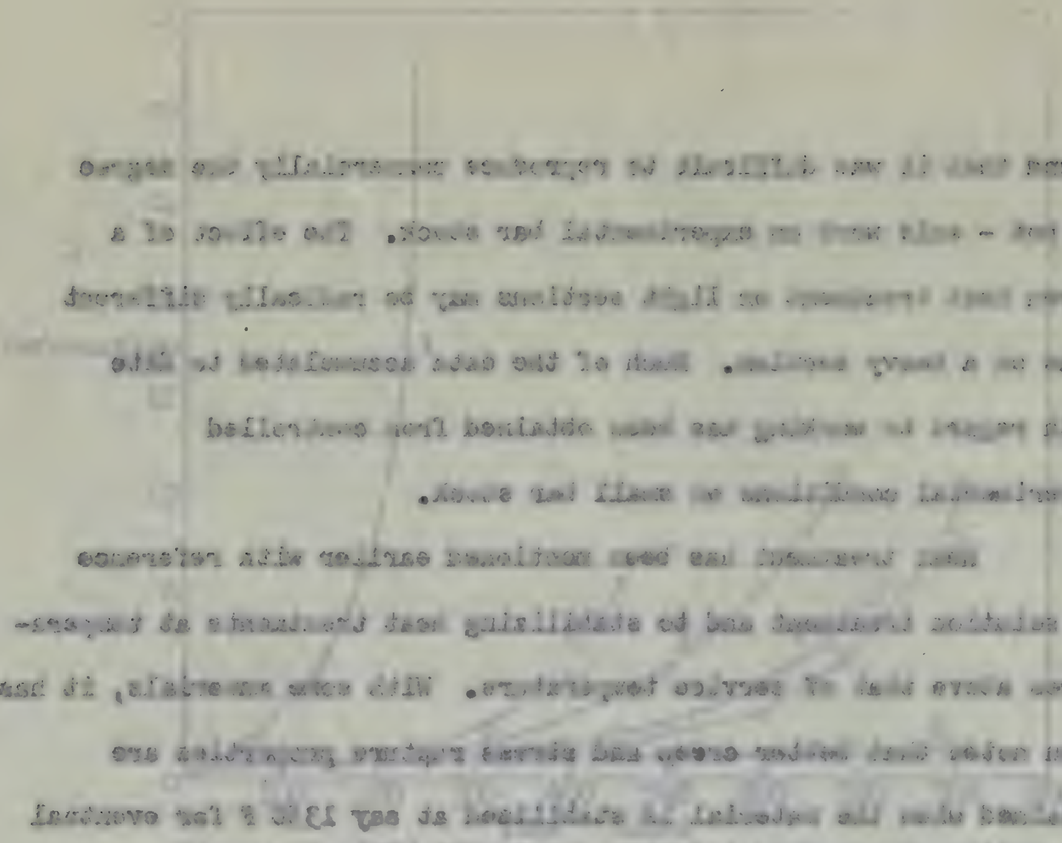


Figure 21. (20) Effect of heat treatment on elongation of N-155 alloy at 1200 F under constant stress of 8000 psi.

Curve No.	Number of heatings to initialforging temp.	Initial Forging Temp.	1st	2nd	3rd	4th	5th	6thing	Finishing temp.after final heat	Time to fracture at 1500F under8000 psi
1	4	1975 2075	30	36	43	55			1300	239
2	3	"	55	43	55				1375 - 1550	490
3	5	"	30	36	36	39	36		1475 - 1525	714
4	5	"	30	36	34	39	36		1300	760
5	1	2050	66						1600	584
6	6	1975 2075	16	15	36	44	30	36	1600	1200

In Figure 21 is illustrated the elongation characteristics of the same series of alloys shown in Figure 20. The heat treatment for 1 hour at 2280 F and air cooling has eliminated the extreme variations observed in Figure 20.

The extent to which strain hardening may be utilized to gain strength is a function of the temperature of operation and the composition of the alloy. At temperatures above the recrystallization or recovery temperature (recrystallization temperature is dependent on time at temperatures and percent deformation) of the matrix, prior strain-hardening of the metal is detrimental as it increases the rate of creep, whereas at temperatures below the recrystallization temperature, strain-hardening processes are helpful. Many of the recently developed high temperature wrought alloys recrystallize near 1350°F. (% deformation not specified)

Grant (33) in the course of preparing Ni-Cr-Co-Fe base alloys, noted that certain grades were totally unforgeable. All of these unforgeable alloys contained tantalum plus columbium from 6.01 to

9.32%. Both of these elements are very strong carbide and nitride formers and it is the presence of these brittle phases in a continuous nature in the austenite matrix which causes failure during forging. Subsequently it was found that alloys containing greater than 4% of either Cb or Ta alone were forgeable. More than 5% of combined Ta plus Cb should be avoided in trying to produce a forgeable product where nitrogen and carbon are present in amounts normally encountered in these alloys. Grant found several other alloys which were unforgeable, but these alloys contained carbon greater than .35% and in certain instances titanium.

At Allegheny Ludlum, surface conditioning of forgings is carefully done between reductions on S-588, S-590, and S-816. To develop the best high temperature properties, the alloys require a double heat treatment. A water quench from 2150 - 2250 F is followed by an aging treatment at 1400 F for a minimum of 10 hours.

Magnaflux, Zyglo, ultrasonic, and X-Ray examinations may be used to insure that forgings are sound.

According to Fonda (112) the most important consideration with regard to turbine discs or wheels is ductility and the best method of obtaining sufficient ductility is by proper control of grain flow during the forging operation. When it was found that the rims of very few wheels reached or exceeded the bursting speed, it was realized that the center rather than the rim was the critical area. It is very difficult to control the "quality" of the center of a large forging.

In rolling S-588, S-590 and S-816 it was found that "hand round" passes which avoid corners and give small reductions in area per pass had to be used. Rolling was done at slow speeds. In machining the above alloys (7) it was necessary to use sulfurized cutting oils, reduced speeds, very sharp cutting tools, and continuous cutting. About the best all-around condition for machining appeared to be a partially - aged state as produced by holding the solution treated stock for 1 to 2 hours at 1400 F. Speeds used in turning were 10-15 feet per minute using high speed tool steel. Drilling is the most difficult operation with these metals. The speed of drilling must be slow. Turbine blades have been machined from rhomboid sections of S-590 in a partially aged condition.

Schmidt (114) has remarked that there are two ways of milling high strength material to obtain reasonable tool life:

1. Reducing cutting speed and feed and cooling the cutter.
2. Increasing the cutting speed and feed, but preheating the workpiece to 1000 F - 1500 F.

Care must be taken to provide protection against heat flow from the workpiece to the machine. Protection can be afforded by insulating with asbestos or by circulating a coolant in the table or fixture.

Chisholm has said (82) that the following factors must be considered when welding high temperature alloys:

1. Ultimate tensile strength and yield strength at room temperature

2. Coefficient of expansion - for most of the high temperature alloys is 50% greater than that of mild steel.

3. Heat conductivity is only $\frac{1}{4}$ that of low carbon steel.

4. Electrical resistance in work hardened condition is greater than that of carbon steel for most of these alloys.

5. Formed and work hardened parts should be annealed before welding.

Blades have been welded to discs and discs have been welded to each other to form the complete rotor. The principal difficulties have been notch extension cracking and weld metal bead cracking. Notch extension cracks are cracks, initiated at the junction between buckets, which propagate in a radial direction across the weld joints. Weld metal bead cracking is longitudinal cracking in the center of the deposited bead, or sporadic intergranular cracking.

Linnert (63) found that combinations of disc and blade material with the least susceptibility to extension cracks showed little difference when welded by submerged arc or by coated electrode. The shape of the submerged - arc deposit appeared to influence combinations of alloys which were sensitive to extension cracking. Wide beads with tapering sides cracked more severely than narrow beads with straight sides. Linnert further found that preheating had no influence on the severity of extension cracking. Weld metal micro structure consisting simply of austenite showed practically

no susceptibility to extension cracking. When small pools of complex carbides and other compounds were present in the structure, an increased susceptibility to extension cracking was present. It was found that the junction between buckets function as minute gaps at the weld joint and reduce the thermal flow from the weld metal opposite these junctions. This slower cooling rate induced some compositions of weld metal to produce pronounced stringer - like segregations extending from the base of the junction notch. Thus joint design is very important. Henry (7) has said that the characteristics of welded joints in these alloys are so dependent on the design of the joint that it is difficult to assemble any basic data of value. Cross (40) and Williams (80) found that interbucket notch extensions occurred even in readily weldable material when the design of the weldment was such that unavoidable notches occur at weld interfaces, transverse to the direction of welding.

Comparison of the composition of weld metal deposit with crack susceptibility is very difficult since composition of the deposited metal could not accurately be determined. Welding electrode materials are available with balanced analysis providing for composition change during welding (89) (84). In many cases, however, electrode materials of dissimilar composition may be used for joints in order to secure stronger welds.

Weldability testing in this field is similar to weldability no susceptibility to subsolidus cracking. When small pools of eutectic carbides and other compounds were present in the structure, an increased susceptibility to subsolidus cracking was present. It was found that the junction between certain [illegible] in [illegible] pass at the weld joint and reduced the thermal flow from the weld metal opposite these junctions. This slower cooling rate induced more compositions of weld metal to produce interdendritic stringer - like segregations extending from the base of the junction notch. Thus joint design is very important. Henry (7) has said that the characteristics of welded joints in these alloys are so dependent on the design of the joint that it is difficult to separate any basic data of value. Jones (41) and Williams (90) found that interdendritic micro fissures occurred even in readily weldable material when the design of the weldment was such that considerable stresses occur at weld interfaces, transverse to the direction of welding.

Selections of the composition of weld metal deposits with crack susceptibility is very difficult since composition of the deposited metal could not accurately be determined. Welding electrode materials are available with balanced analysis providing for composition changes during welding (90) (91). In many cases, however, electrode materials of dissimilar composition may be used for joints in order to secure stronger welds.

Weldability testing in this field is similar to weldability testing in general. There are as many different tests as there are applications. Harris and Oliver (35) have outlined a test procedure which they applied to flash butt-welded discs and shafts. The criterion of a successful weld was whether the tensile strength across the weld was close to the tensile strength of the shaft material (which was the weaker member of the disc - shaft pair) and whether the fatigue resistance of the weld junction was close to that of the shaft. Microscopic examination of the weld junction was also done at 20X. Any weld junction which showed more than a certain amount of oxide film at the weld was rejected. In addition a torsion test was run on the material in order to see if cracks formed on twisting under a certain stress.

Larige (21) has mentioned making test pieces, conforming as nearly as possible to the weld in question, and then metallographically and radiographically inspecting them. The test pieces were given tensile tests in order to determine the weld strength. Thus most of the weldability tests are a modified service test. Data and results from one of these tests may not be applied generally since, as has been seen, joint design, material, and electrode material are controlling variables.

testing in general. There are as many different tests as there are applications. Morris and Wilson (45) have outlined a test procedure which they applied to flash butt-welded discs and shafts. The criterion of a successful weld was whether the tensile strength across the weld was close to the tensile strength of the shaft material (which was the weaker member of the disc - shaft pair) and whether the fatigue resistance of the weld junction was close to that of the shaft. Microscopic examination of the weld junction was also done at 50X. Any weld junction which showed more than a certain amount of oxide film at the weld was rejected. In addition a torsion test was run on the material in order to see if cracks formed on twisting under a certain stress.

Lawson (43) has mentioned making test pieces, conforming as nearly as possible to the weld in question, and then metallurgically and radiographically inspecting them. The test pieces were given tensile tests in order to determine the weld strength. Thus most of the weldability tests are a modified service test. Data and results from one of these tests may not be applied generally since, as has been seen, joint design, material, and electrode material are confounding variables.

CHAPTER V

SHORT TIME TENSILE TEST

Many investigators have said that the tensile test is valueless as a measure of long time performance at high temperature. (115, 60, 29). Freeman and others, working with Timken 16-25-6, low carbon N-155, S-590, S-816, and inconel X found no consistent relationship between tensile properties and rupture, creep, or stress-time for total deformation characteristics. Gillett has said that short-time high temperature yield strength tests reveal nothing about creep behavior. Under conditions of stress and temperature in the tensile test where remnants of strain-hardening remain, metals do not "settle down" to a consistent creep behavior(27). On the other hand Evans (30) and Henry (7) have said that the hot tensile test does furnish useful information for hot working or hot forming operations. For example warning of temperature zones in which the alloy has a tendency to become brittle may be obtained by points or data which are radically removed from the contour of plotted curves. Bobrowsky (14) found an approximate correlation between modulus of rupture and short-time tensile strength based on results for brittle materials(ceramals) at a number of laboratories. Modulus of rupture is defined as the calculated maximum stress in the outer fibers of the specimen at fracture. The short-time tensile strength of the specimens (TiC with varying proportions of Co, W, or Mo) was found to be about 0.4 to 0.6 of the modulus of rupture.

In any discussion of tensile testing, the modulus of elasticity is brought to mind. Mack (99) has listed the following points dealing with the modulus:

1. For any element, the modulus is a function of atomic volume, melting point, and the temperature.
2. Crystallographic effects seem very important because of elastic anisotropy.
3. The modulus decreases with increasing temperature.
4. Cold - working reduces the modulus while stress - relieving and process anneals tend to increase it.

Avery and Wilks (48), working with cast 25 Cr-20 Ni alloys, found that in the conventional tensile test at high temperature the value of the modulus was considerably affected by the speed of testing. Cross and Simmons (47) and Badger and Sweeney (45) working with vitallium type alloys found a variation in the modulus of elasticity with the grain size. In the table below (45) note the narrowing in the variation from maximum to minimum as the grain size decreases. This is explained on the basis of grain orientations. With a fewer number of grains(larger grain size), there is a possibility for greater variation in tensile properties since each grain is responsible for a greater percentage of the total strength

Actual Average Grain Diameter	A.S.T.M. G.S.	Modulus of elasticity, psi Max.	Min.	Average
0.003	5	36,900,000	35,470,000	35,960,000
0.012	1	35,200,000	31,250,000	33,120,000
0.050		40,600,000	23,400,000	31,700,000

of the cross section, whereas with a fine grained specimen, each grain is responsible for only a small part of the total strength of the cross section.

Roberts and Nortcliffe (91) have used a dynamic method for measuring the modulus at high temperatures. In measuring the modulus by usual static methods with the tensile test at high temperatures, great difficulties are encountered since only small stresses may be used if creep in the specimen is to be avoided. The combination of using very small stresses with consequently small strains leads to the possibility of making experimental errors of large magnitude. In the dynamic method, a cylindrical bar of the metal is caused to vibrate transversely with its fundamental frequency. From the frequency and the mass and dimensions of the bar, Young's modulus may be calculated (formula not given). The results showed a decrease in Young's modulus with temperature. The value at 1100 F for 3C - 2.25 Ni - .39 Cr - .4 Mo was 3/4 of the modulus at room temperature.

Hardness may be correlated with tensile strength of steels at room temperature. Evans (30) has said that hot hardness is not needed for most high temperature applications (exception valves). Besides, this property does not correlate with the other more important properties such as stress rupture and creep. Parke and (44), however, in making a selection of promising materials for use as gas turbine blades for temperatures up to 1600 F used hot hardness as a

primary test on the basis that the harder materials had greater creep strength.

In tension testing at elevated temperatures the following points must be considered: (126)

1. Method of heating specimen
2. Measurement of specimen temperature
3. Measurement of extension
    a. elastic range extensometer
    b. plastic range extensometer
4. Measurement of load at fracture

Most laboratories at the present time do not get the yield strength in short time tensile tests at elevated temperatures because they are not equipped with extensometers for high temperatures (21)

In the static method of hardness testing at elevated temperatures, a Vickers hardness machine employs the regular loading device. A vacuum chamber protects the test specimen, the diamond indenter, and the heating element. The impression in the specimen is read by means of an optical system. The apparatus may be used up to 1700F (125). Nisbet (discussion to 125) has said that if the rupture test data are plotted on linear coordinates rather than log - log, the rupture curve falls very fast, with high loads and short times and becomes very flat with low loads and long times. It would therefore by very difficult to predict the horizontal part of the stress-rupture curve from short time data which fall on the almost vertical part of the curve.

Instead of a slowly applied load, the dynamic hardness testing method uses a falling weight, a released spring, or a blow by a

pendulum or hammer to produce the indentation of the test piece (124). The hot working stresses in forging, cogging, rolling, extruding, drawing, and pressing are many times faster than those applied in the static tensile and hardness tests. The differences in static and dynamic hardness versus temperature curves shows that the static method does not necessarily tell the correct resistance to plastic forming at rapid deformation speeds. In extreme cases, the maximum of dynamic hardness coincides with a minimum of static hardness. By gradually increasing the deformation speed in impact torsion and high-speed tension tests, the maximum yield strength and the ultimate strength of iron and steel are shifted to higher temperatures. Thus in order to get a qualitative estimate of conditions which may be expected in hot working the dynamic hardness test at elevated temperatures is more appropriate than the static hardness test.

CHAPTER VI

FATIGUE

Fatigue strength does not appear to be an important issue at high temperatures (63) (27) (17) (11). Gillett (27) has said that in cases in which a long service life is inherent in the design the loads which produce the low limiting creep rates and total deformation are considerably lower than the endurance limit. Therefore creep rather than fatigue is the controlling design factor. This relative unimportance of fatigue strength may be due to a decrease in the notch sensitivity of the steels with rising temperature.

In those cases in which a very limited service life is the basis for design, resulting in high loads and appreciable allowable deformation, stresses which produce fracture in 1000 hours are sometimes of the same order of magnitude as the endurance limits and therefore fatigue must be considered. It is not known how much more important surface condition becomes when fatigue stresses are the controlling stresses in design, nor is there any measure of the role played by alloy composition in connection with surface sensitivity.

In many high temperature fatigue tests, there is no apparent tendency for stress - cycle graphs to level off and develop an apparent endurance limit (27) (30) (11).

With reference again to the decision as to whether fatigue strength or creep is the most important variable— if the fatigue strength and yield strength determined by various mean static stresses

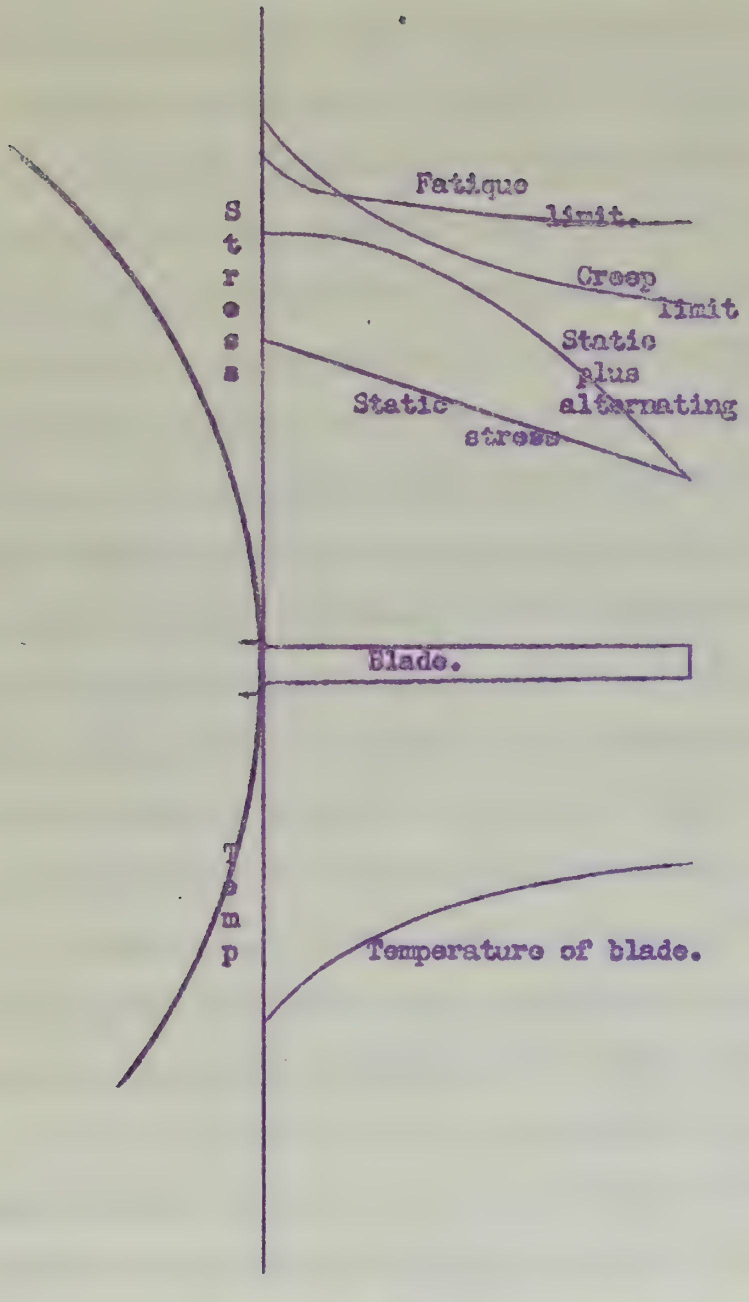


Figure 22. Temperature and stress distribution in a turbine blade. (17).

is determined at various temperatures, it is found that at low temperatures the fatigue strength is always lower than the yield point and at high temperatures the fatigue strength is higher than the creep limit. (17) There is an intermediate temperature at which the static yield point and the fatigue strength are identical. This temperature varies with the definition of the creep limit and the frequency of the load reversals applied when determining the fatigue strength. This temperature indicates the transition from brittle fatigue fracture to extension by creep. Therefore the engineer has an indication up to what temperature fatigue strength is the criterion of the maximum permissible stress, and beyond this temperature where creep limit is the criterion.

In Figure 22 is shown the state of stress in a blade at various positions of the blade compared with the temperature, fatigue limit and creep limit. The outer part of the blade is almost the same temperature as the impinging gas; the root of the blade may be about 100 - 200 F cooler. The creep strength at the root is higher, but the fatigue strength may be lower than at the blade tip. The combined static and alternating stresses are highest at the root. Therefore it is possible to have a situation where the permissible stress is lower in the colder section of the blade. If transverse vibrations of very high frequency arise, the highest stresses will then be at the tip of the blades.

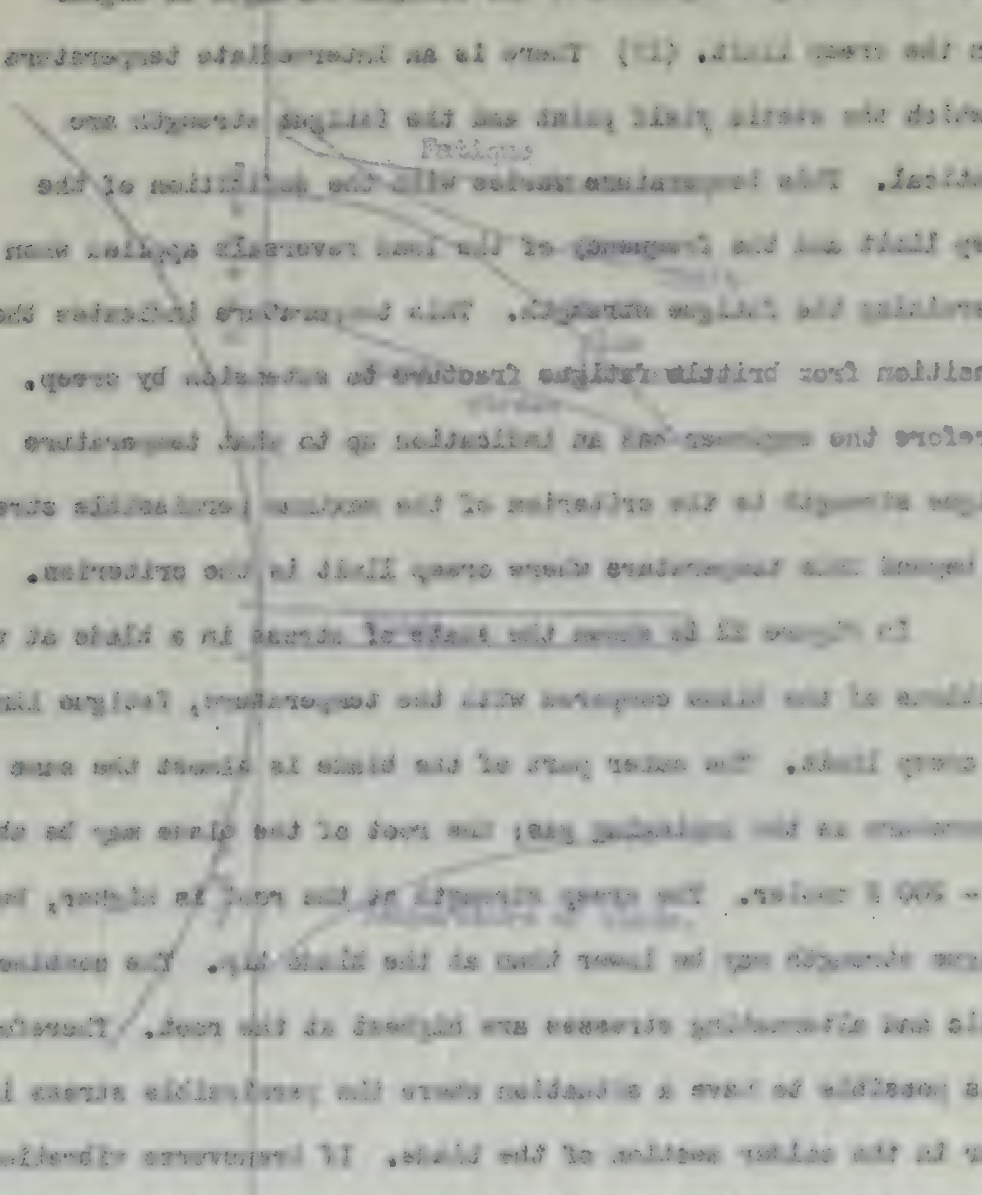


Figure 22. Temperature and stress distribution in a turbine blade. (17)

Turbine blading is required to withstand a steady centrifugal stress, a steady gas bending stress, and complex fluctuating stresses. These fluctuating stresses are the result of aerodynamic buffeting of the moving blades from the preceding row of stationary blades which may set the blades into vibration with a peak movement at resonant conditions when each moving blade receives successive blows from each stationary blade. A flutter phenomena may also occur at the trailing edges (4).

In order for a fatigue testing machine to imitate the stressing of a turbine blade, it would be required to exert simultaneously a uniform tensile load, a uniform bend load, and a flustuating bend load at frequencies of several thousand per second at temperatures exceeding 1200F (129). It is usual to design on the basis of only steady stresses using the elastic theory and creep data.

Since there is insufficient information available for the designer to base his design on fatigue stresses, he insures that the possibility of fatigue failure is made remote at the expense of lightness and aerodynamic efficiency. This explains why no special effort has been made to develop new turbine alloys of which the prime function is the ability to withstand fatigue. Alloy manufacturers feel that a sufficient amount of fatigue resistance accompanies materials of high creep strength (129).

Most fatigue testing simply indicates one alloy to be more resistant than another to vibration in general. The current methods of fatigue testing can be divided into two main classifications. The first is the vibration test in which the main objective is to find the natural frequency of the finished component. The second method is an attempt to establish the endurance limit of the material when subjected to conditions resembling a service test.

Various fatigue tests in use now are:(130)(131)(30)(129)(121)

1. The pneumatic excitation test - vibration test
2. Violin bow - resonance test
3. Alternate tension - compression test. Haigh or Krause machine is used for this test.
4. Wohler Test
5. Westinghouse Test - a high temperature fatigue machine which employs a fixed cantilever, constant deflection principle. Vibrations are induced by a balanced electronic system.
6. Schenk type tests

Edwards (129) has said "Suppose we are asked to determine the maximum fatigue stress permissible for a material to be used in a turbine blade required to last "t" hours in service under a given steady creep stress of "W" tons. The frequency of vibration in the turbine is n cycles per minute whereas the highest frequency possible in the fatigue machine may be $\frac{\text{"n"}}{50}$. There is a choice here of two test procedures. A steady stress of "W" tons may be imposed on the specimen and an attempt to find the limiting fatigue stress for a life of "t" hours may be made. This is the correct method on the basis of time alone. On the basis of stress cycles the correct

limiting fatigue stress would be based on 50 "t" hours." In the first instance, the specimen is only given 1/50 th the reversals it would undergo in service. In the second instance, the specimen is subjected to a creep stress "W" for 50 times longer than it would be in service. Neither method is correct but it is thought that the first method is closer to the correct answer.

With high rates of stressing applying in fatigue testing, a dynamic method of calibrating the equipment in order to determine the stress-strain relationship is necessary. The dynamic limit of proportionality has been shown to be much higher than the static limit in some materials at elevated temperatures. At sufficiently elevated temperatures, there may be no limit of proportionality under static testing conditions and the relationship of stress to strain at high rates of stressing can be determined only by dynamic calibration. The method of calibration described by McKeown and Back (130) makes it possible to decide in any given instance whether or not it is justifiable to assume elastic stressing in computing bending stresses from the applied force and specimen dimensions. For example, when the shape of the stress - strain curve or load - deflection curve is affected by the rate of stressing at elevated temperatures, it is necessary to determine the curve at a rate of stressing the same as that occurring in the fatigue test. Creep occurring during a calibration at a slow rate of stressing, referred to as a static calibration, may

give a non-linear stress-strain relationship whereas, if time for creep were not available because of the high rate of stressing, the relationship may be linear up to much higher stresses.

With reference to the fatigue strength of particular alloys, Toolin and Mochel (132) plotted 70 fatigue curves at 1200 and 1500F on several high temperature alloys. At 100 million cycles, the fatigue curves of most of these alloys approached a straight line on the standard semilogarithmic stress - cycle diagram. The slope of this straight line section of the diagram varied from alloy to alloy. The wrought form of the alloy appeared to give higher fatigue strength at 1200 and 1500 F than the precision cast form. Fatigue fractures of the wrought materials were, in general, transcrystalline. Alloys with higher before - test hardness showed superior endurance properties.

Evans (39) found that with a 12 Cr alloy at 1000 F stress cycle curve was still falling at 500,000,000 cycles (for the frequency used - 1158 hours).

Cross (40), using a Krouse direct - stress machine with a stress amplitude of ± 15,000 psi at 1350 F on forged S-590 and N-155 (low carbon), found ductile fractures with measurable elongation and necking down. The time to rupture in the fatigue test closely approximated the constant-stress to rupture time. This would indicate at least for this particular experiment that cyclic stress had no effect on the time to rupture.

With reference to the previous statement made regarding the possibility of more severe stress conditions at the root than at the blade tips, Bobrowsky (14) successfully operated a rotor with 3 ceramal blades and 139 alloy blades at gas temperatures up to 2000 F and speeds up to 15000 rpm. Almost simultaneously with an increase of speed to 17,500 rpm., all of the ceramal blades failed at the root where the temperature was quite low.

Where damping is accomplished by the mechanical blade-disc joint or by the inherent high damping capacity of the material itself, fatigue obviously is not a worrisome problem. However, the trend to welded joints and the fact that austenitic alloys (and most of the present high temperature alloys have an austenitic matrix) points to problems in fatigue since welded joints and austenitic alloys have poor damping characteristics.

Wilkes (120) has found certain changes occurring in the internal damping of S-816, Inconel X, N-155, and Timken 16 Cr - 25 Ni - 6 Mo as a result of continuous vibration. For example at a peak stress of 40,000 psi, the Ni and Co - base alloys showed a minimum damping near 800 F and a sharp rise between 1350 - 1500 F. The two iron-base alloys had maximum damping between 800 - 1200 F, a slight dip at 1350 F, and a sharp rise at 1500 F. N-155 and S-816 indicated excessively high damping at 1200 F and above at stresses below its fatigue limit. This high value however became rapidly smaller as vibration was continued

to the extent that it finally showed lowest values of the four materials. This means that damping data obtained by more orthodox methods in short times may, in some cases, be greatly in error if used in choosing materials for gas turbine buckets. Consequently trouble would result in those designs where internal damping of the buckets is a major factor in limiting vibration.

Finally it may be said that the damping capacity of a material appears to be an inherent characteristic of the material. It is relatively unaffected by heat treatment or other processing variables. Ferritic materials have high damping capacity. 12 Cr steel has one of the highest damping capacities known. (30)

GENERAL

THERMAL CONDUCTIVITY

A high thermal conductivity is a requisite in materials expected to possess high thermal shock resistance. With materials of high thermal conductivity, it is possible to operate at higher gas temperatures than with materials of low thermal conductivity - other conditions being the same. An experiment in which high temperature alloy (metal), ceramic, and ceramal blades were mounted in a wheel and then passed through the flame from a gas torch for a given time revealed that the ceramic became white hot, the alloy bright red, and the ceramal a dull red (14).

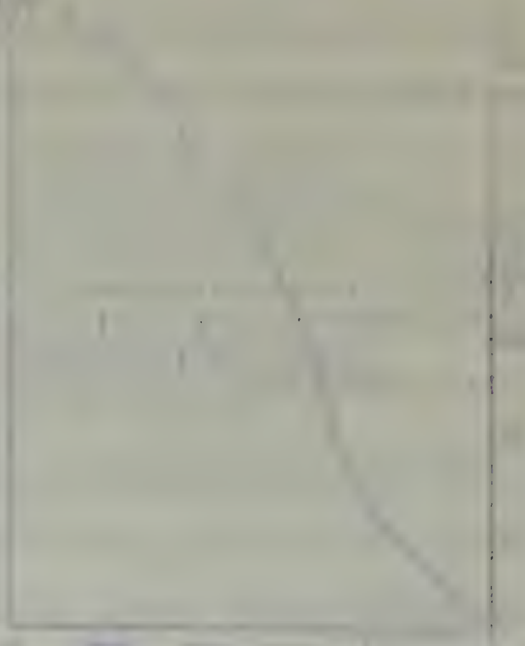
COEFFICIENT OF EXPANSION

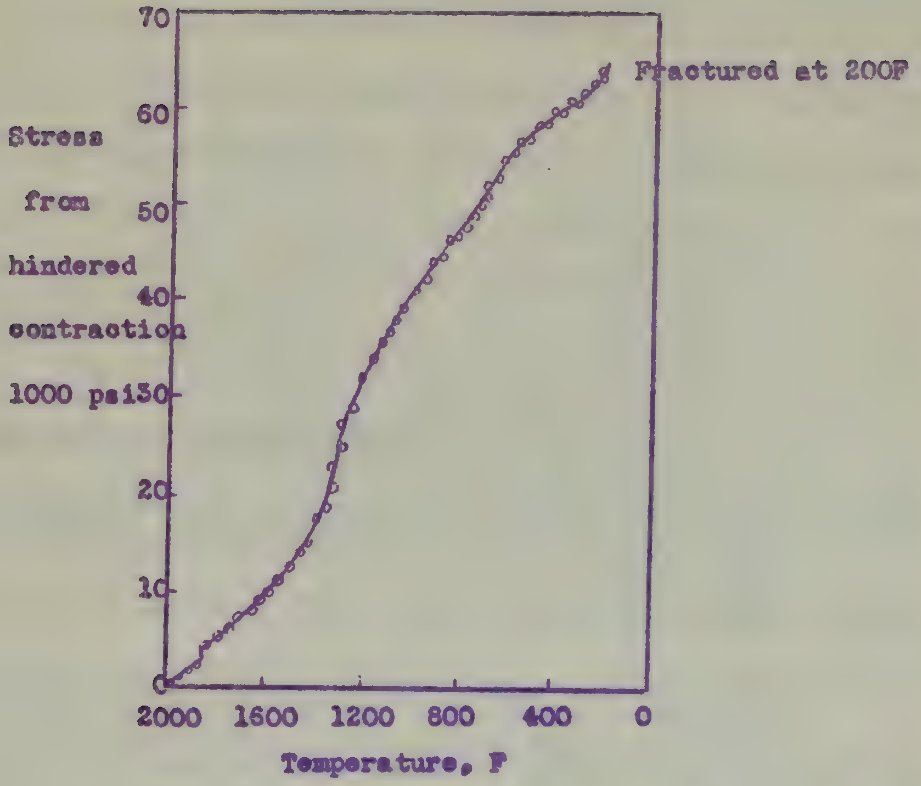
A low coefficient of expansion is desirable where close tolerances and cyclic operations are necessary. A low coefficient of expansion has much to do with the adherence of oxide films. In welding a low coefficient is desirable. Unfortunately many of the high temperature alloys have coefficients as high as 1.5 times that of the plain carbon steel. The coefficients of these alloys increase with temperatures. For example, Cunningham has reported (24) that the Elliott turbine is nearly 1/2 inch longer when it is operating than when it is cold. Due to this expansion at the high temperatures of operation the turbine must be balanced not only at speed but also under conditions of temperature. (Normal method of balancing turbines for operation at moderate temperatures 500 - 600 F is a dynamic

balance at room temperature.) In the Elliott turbine expansion is handled by having the turbine and its appendage fixed at its exhaust end and with all expansion forced toward the inlet, where connections float by freely moving links.

The specific gravity or density of the material influences the magnitude of the stresses developed in the rotating buckets due to centrifugal force.

Parke and (44) in the search for a suitable metal for operating temperatures up to 1600 F used the melting point of the metal as the criterion for selection. It is noted that, although pure metals have higher melting points than their alloys, the pure metals do not usually have as good mechanical properties. Tungsten (M. P. 6190 ± o6 F) seemed to be the ideal base. Unfortunately, tungsten - chromium alloys with excellent oxidation resistance and high strength at 1600 F were too fragile and difficult to fabricate.





Alloy	C	Mn	Si	Ni	Cr	N
HK-20a	.20	.92	1.36	21.0	24.2	.154

Initial stress at 2000F psi	Cooling rate (1200-1000F) F/min.	Total R.T. Elong. %	Total R.T. R.A. %
350	22	17	20.9

Figure 23. Plot of hindered contraction , stresses versus temperature. (48)

THERMAL SHOCK RESISTANCE

Materials for use in gas turbines specifically designed for transportation which implies cyclic operation must have a certain amount of resistance to thermal shock. Norton (15) has summed up the properties required by materials at elevated temperatures for maximum resistance to thermal shock as:

1. High tensile stress
2. High thermal conductivity
3. Low specific heat on a volume basis
4. Low effective modulus of elasticity (ratio of stress to strain at fracture)
5. Low coefficient of thermal expansion
6. Low emissivity
7. Low exterior conductivity (heat loss by convection at the surface)

Resistance to failure from thermal shock was determined by Bobrowsky (14) by heating circular disks 2 inches in diameter and about ¼ inch thick to a given temperature and then quenching these disks in a moving stream of air. If a specimen withstood 25 cycles of shock at a given temperature, the test was continued at the next higher temperature for another 25 cycles to a maximum temperature of 2400 F. The thermal shock parameter was set equal to

$$\frac{(K)\ (T.S.)}{(E)}$$

where K thermal conductivity
T.S. tensile strength
coefficient of thermal expansion
E effective modulus of elasticity

A large value of the above parameter is indicative of high resistance to thermal shock.

Avery and Wilks (48) have plotted the stresses which result if the thermal contraction of an alloy is hindered. The plot shown in Figure 23 may serve two functions: (1) It may indicate the magnitude of thermal stresses that can develop. (2) It provides a quick criterion for determining at what temperature heat treatment must be conducted to reduce residual stresses below a certain level. Above the corresponding temperature point on the graph, any stress that is applied to an alloy will rapidly fall by plastic deformation to the value indicated by the curve.

Henry (11) has remarked that substitution of fine grained material for coarser grained material permitted relief of thermally induced stresses by local elongation at points of stress concentration.

NOTCH RESISTANCE

Siegfried (58) has said that heat resistant alloy steels tested at high temperatures in notch impact behave like ordinary basic Bessemer steel tested at low temperatures. Small cracks in the surface may slowly develop and lead to fracture without deformation. Schaub (51) has remarked that the limiting creep stress should be complemented by a corresponding test (creep) on notched test bars. In the notched test a stable material should not exhibit any tendency to embrittlement or to fissure formation in the notch similar to stress corrosion.

CHAPTER VII

CORROSION AND OXIDATION

There are a number of reasons why a clear cut corrosion or oxidation index of a material cannot be given. The rate of corrosion and oxidation of a material depends on many different variables. These variables are:

1. Temperature
2. Ambient atmosphere
3. Composition of material
4. Time of exposure
5. Thermal shock conditions - coefficient of expansion
6. Mechanical stress and abrasion
7. Steady or fluctuating temperature and composition of atmosphere
8. Method of reporting corrosion loss or oxidation

Gow (99) has said: "Broad generalizations regarding the relative corrosion resistance of different alloy compositions based on the experience gained in one type of application or on the results of tests conducted under one set of conditions are likely to be wrong in their implications when corrosion conditions other than those for which the observations were made are encountered."

Harwood and Schulman (101) have made a rather comprehensive report on current corrosion research which is being conducted under the auspices of the U. S. Navy. Oxidation and tarnishing are now recognized as electrochemical phenomena. Metal ions have been shown to diffuse

outward through the oxide (sulfide or halide) film as well as the less frequent diffusion of reactant (gaseous) ions inward. A growing film may be compared to a current - producing cell with the metal-film interface (as anode) supplying cations and electrons for outward diffusion and with the attacking substance - film interface (as cathode) supplying the anions for inward diffusion. The film acts as both the internal and external circuit of a closed cell.

Metals can be divided into two groups for consideration of their oxidation products: (101)

1. Lighter metals form porous oxides which have a smaller volume than the equivalent metal consumed in producing these oxides. These metals generally exhibit a linear rate of oxidation at constant temperature.
2. Aluminum and the heavier metals form protective, non-porous oxides of greater volume than the equivalent metal consumed. Many metals of this group exhibit a parabolic oxidation rate (inversely proportional to the square root of the time) Under some conditions the oxidation process is proportional to the logarithm of time.

Fundamental studies of oxidation and corrosion are being made currently at several universities. At Illinois Tech (101) direct measurement of the rates of gas - metal reactions is being made by means of a gas volumetric method with gas circulation under atmospheric or low pressure. One of the initial phases was the study of electron potentials in growing halide and oxide layers. A study of the zone of metal phase consumption (locus between the initial surface of the

metal and the surface of the

retreating metal phase) revealed that even in the cases where a scale grows by outward diffusion of the metal or metal ions, the scale also must grow on the other side towards the metal. Conditions of scale growth on both sides of the metal are dissimilar. Thus, the structure of the scale tends to be different in both layers formed. Composition and structure of films are being investigated by chemical and metallographic examination and by electron diffraction studies. Several stainless steels have been exposed to varying oxygen atmospheres for different temperatures and time treatments. The oxidation process has been followed by weight measurements and diffraction studies. It was found that the oxide films formed in low oxygen atmospheres at 1800 F are more protective to further oxidation than scales formed in higher - oxygen atmospheres. Electron diffraction patterns have been obtained at the temperature of oxidation.

The high temperature oxidation products formed in oxygen and in air in Cr steels, ranging from 5-26% Cr content, have been analysed by X-Ray diffraction. A regular variation of phases present can be correlated with change in oxidation resistance. In contrast with pure iron, no FeO was found in the oxidation of iron - Cr alloys indicating a possible function of Cr in producing an oxidation resistant scale. Since it is recognized that the protection afforded a metal may not be due to the bulk oxide acting as a mechanical barrier, but rather to some special properties of a relatively thin layer next to the metal, film stripping techniques are being utilized to examine thin films by transmission electron diffraction.

At Carnegie Institute of Technology (101), the diffusion mechanisms operating during oxidation processes are being studied by means of radioactive tracers. It is possible to determine if it is iron or oxygen which diffuses through the oxide layer during the scaling of iron. In brief the method is of electroplating a thin layer of radioactive iron on a polished surface of pure iron. The specimen is then oxidized under controlled conditions. The distribution of radioactive iron through the scale indicates the nature of the diffusion process.

At Stanford University (101), a program is underway to determine if atmospheric reactions have an important effect on stress - rupture properties of high temperature alloys. Then by the use of pure metals, the mechanism and nature of these effects are to be studied. The results to date have indicated that atmospheric reactions have a definite effect upon high temperature strength properties of metals and alloys. The program will also include the study of the effect of strain rate on the embrittlement of pure metals subjected to stress rupture tests in active and inert atmospheres at elevated temperatures in order to determine if the reduction in ductility is due to environment or effect of low strain rates and high temperatures.

From the above discussion, it would appear that very little fundamental study has been done in the field of corrosion. This is not so however. Hickman (95) (96) and Hickman and Gulbransen (94)

have reported a number of electron diffraction studies on stainless steels 301, 446, S-588, nichrome, inconel, stellite, refractaloy, hipernik, kovar, and hastelloy. Hickman (95) used the reflection electron diffraction technique in studying the structures of the oxide films on the above metals at elevated temperatures. The observations were made at elevated temperatures and after the metal had been cooled from the high temperature. The following results were noted:

(1) The oxide which formed on most of the alloys was Cr_2O_3.

(2) In general, oxides of iron are observed at low temperatures while oxides containing Cr appear as the temperature is elevated.

(3) Heating and cooling experiments show that solid phase reactions occur during temperature cycling.

(4) There does not appear to be any correlation between the structure of the alloy substrate and the oxides which may form on its surface. Both austenitic and ferritic alloys form Cr_2O_3 and spinel type oxides.

(5) On those alloys which contain Ni and Cr (nichromes) and those which contain Co and Cr (refractalloy and stellite), oxides of Ni and Co

are observed at low temperatures while Cr oxides appear as the temperature is raised.

Hickman and Gulbransen (94) found that, if the oxide coating of these alloys was protective to the alloy at temperatures in excess of 1300 F, the structure of this protective coating is of the spinel type.

Hickman (96) has mentioned a number of factors important in determining the chemical and physical structure of the oxide film on the surface of an alloy. Some of these factors not previously mentioned are:

1. Thermodynamic stability of the oxide
2. Lattice type of the oxide and its resemblance to the metal or alloy lattice.
3. Preoxidation treatment of the metal such as annealing, polishing, and cleaning.
4. Rates of nucleation and growth of the oxide crystals.

"4" above is unknown, "3" has been estimated from thermodynamic data, and "2" has been determined from diffraction studies. Some work along preoxidation treatment lines has been done by Vernon, Wormwell, and Nurse (97). The relationship of the lattice type of the oxide to that of the underlying metal or alloy may determine which oxide will form. It is possible that a given oxide may have greater probability of forming if its lattice matches that of the metallic substrate. Goldschmidt (98) has shown that FeO and Fe_3O_4 have lattices which may be readily derived from the lattice of alpha iron, while alpha Fe_2O_3

has a rhombohedral lattice which differs considerably from that of the body - centered cubic alpha iron. Experimentally Fe_3O_4 and FeO are found in contact with alpha iron while alpha Fe_2O_3 is not. Binary alloys present added difficulties in comparison with the metals since the oxides which form may be complex and contain the two metals or solid solutions of various oxides of the two metals. Investigations of binary alloys have shown that it is not possible to predict the oxide that will form on the surface on the basis of the thermodynamic stabilities of the several possible oxides. Oxides of some of the elements, silicon and zirconium, are not observed in the outer surface of the oxide layer. Oxides of one metal (in a binary alloy) occur at low temperatures, while oxides of the other metal appear at higher temperatures. The problem of identifying an oxide on the surface of an alloy may be extremely difficult if two or more of the oxides below are present simultaneously. The reason for this difficulty is the similarity in structure and unit cell sizes which results in the same distribution of diffraction lines and intensities of these lines. The similarity in structure also increases the probability of solid solution of one oxide in another.

Oxide	a(A)
$CoO \cdot Co_2O_3$	8.11
$FeO \cdot Al_2O_3$	8.12
$NiO \cdot Cr_2O_3$	8.31
$CoO \cdot Cr_2O_3$	8.32
$NiO \cdot Fe_2O_3$	8.34
$FeO \cdot Cr_2O_3$	8.32
etc.	

Correlation of oxidation rate with the constitution of the oxide coating should be made in order to determine which additions should be made to the base material for best oxidation resistance (14).

and McDonald (100) have said that temperature increases the rate of scaling 1.2 to 2.0 times per 18°F temperature increase providing the temperature differences are not too large and the same oxide phase is being formed.

The effect of the pressure of oxygen, in the case of oxidation, acts differently in various pressure ranges. At very low oxygen pressure, the dissociation pressure of the oxide scale may be higher than the external oxygen pressure. Under these conditions scale does not form. When the external pressure exceeds the dissociation pressure, scale forms. If an oxide is present in a closed system without an additional supply of oxygen, and another metal is placed in the same closed system, then the oxide may dissociate giving off oxygen and the second metal may oxidize providing the conditions described above obtain.

The corrosion behavior to be expected of an alloy cannot be predicted from known corrosion behavior of component elements and the proportions of these elements in the alloy (99). In Fe-Ni-Cr alloys, the optimum Ni or Cr content for maximum corrosion resistance varies as the Ni or Cr levels being considered are altered. Also the optimum

amount of Ni and Cr varies with the temperature being considered. Co, W, and Mo which have relatively poor oxidation resistance may actually enhance the oxidation scaling resistance when present in particular amounts in Fe - Cr - Ni alloys (99). However, if volatile oxidation products are formed such as WO_3 or MoO_3 the alloy may have lower oxidation resistance as a result of the addition of the tungsten or molybdenum (103).

Perhaps the most clear cut statement that one can make about the oxidation of high temperature alloys is that the resistance to oxidation is a function of the chromium content. Since most of the high temperature alloys have an austenitic matrix, an increase in chromium content as operating temperature increases necessitates that nickel or some other austenite stabilizer be increased to maintain the austenitic structure. Gow, Brasunas, and Harder (43) made corrosion tests of Fe - Ni - Cr alloys in air at 1600 to 2200 F. They found that Cr content is considerably more effective than Ni content in resisting oxidation at 1600 F. Two general composition ranges where optimum corrosion resistance was obtained at 1600 F and 2200 F were 26 Cr-20 Ni and 16 Cr - 60 Ni.

Lincoln (28) has listed the following percentages of Cr as being necessary in Cr - Ni steels to resist excessive oxidation at the indicated temperatures:

Cr	Temperature
12%	1300 F
16%	1500 F
27%	2150 F

In addition to oxidation in general, decarburization, carburization, nitridation, interdendritic or intergranular penetration, and selective infiltration attack may occur. The presence of carbon in combustion gases as unburned hydrocarbons or in other forms may cause the carbon to diffuse into the steel at an appreciable rate at temperatures in excess of 1300 F. The unfortunate thing about carbon in a chromium steel is that carbon combines with sixteen times its weight of Cr to form chromium carbides. Thus in the vicinity of these carbides the matrix may be depleted below the amount of chromium needed for oxidation resistance (establishment and maintenance of chrordium oxide scale) at the particular temperature. Lincoln added 1.7 Si to a 25 Cr-20 Ni steel and then exposed the steel to a carburizing atmosphere for 160 hours with practically no carburization. For the optimum resistance to carburization, straight chromium steels which are completely ferritic are recommended.

Sulfur present in combustion gases either in the elemental form or as H_2S attacks steels with nickel content. The sulfur forms with the nickel a Ni-S eutectic which melts at 1193 F. Lincoln(28) has found that in atmospheres which are oxidising and in which the sulfur remains as SO_2 there is slight attack on straight chromium steels or steels with low Ni content. Under reducing conditions with sulfur in the form of H_2S, there is serious attack at elevated

temperatures particularly if the steel contains nickel. Stauffer and Kleiber (92) (102) found that in some cases, high sulfur combustion gases caused less scaling than low sulfur gases. They attributed this to the great scatter in the test results.

One of the discouraging things about expressing a corrosion index of a metal aside from meeting the controlled conditions of time, atmosphere, and temperature already mentioned is that the weight lost method is subject to great experimental error. For example, Gow, Brasunas, and Harder (43) could not get a completely metallic surface on Fe-Ni-Cr alloys after repeated and prolonged descaling. McCullough (93) has said that descaling (associated with weight lost) is not satisfactory for long-period tests where descaling will not indicate subsurface destruction due to intergranular attack and decarburization. Nevertheless, corrosion data is expressed as "grams lost per inch^2 per day" or "inches penetration per year." The German criterion for corrosion resistance of steel for gas turbine blades during World War II was that at the given temperature the tested steel had to show a loss not greater than 1 gram per hour per square meter, and at 90 F above that temperature a loss not greater than 2 grams per hour per square meter. The above figures appear to be very arbitrary and at the best probably have very restricted engineering applicability.

CHAPTER VIII

TESTING

The mechanical testing of the high temperature alloys for dynamic applications only emphasizes how little information is actually obtained from such tests as creep, tensile, and fatigue with relation to performance in service. Sweeney (21) has said that the general belief today is that laboratory mechanical test data only serve as indications of the uses for which an alloy appears to be suitable. At the present time, the final choice of material is dependent only upon long-time service testing in the turbine unit. As an example of the lack of agreement on any one alloy for a specific application, Sweeney has said that of the five cobalt base alloys which Haynes Stellite Company makes for precision - cast turbine blades Allison thinks number 21 alloy is best, Westinghouse has always used number 23, and other turbine builders feel that alloys number 27, 30, or 31 are better.

Badger (25) has described the service testing of wheel materials in which the wheels are unbucketed to failure in a high-speed vacuum stand. About 150 wheels were burst to determine the relative quality of various manufacturer's wheels and to evaluate X-Ray, zyglo, and supersonic testing. In general, it was found that X-Rays were effective in locating internal pipe or blowholes but

CHAPTER XII

SERVICE

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Badger (25) has described the service testing of steel materials in which the wheels are   subjected to failures in a high-speed rotor stand. About 150 wheels were burst to determine the relative quality of various manufacturers' steels and to evaluate X-ray, sonic, and magnaflux testing. In general, it was found that X-rays were effective in locating internal pipe or discontinuities but ineffective in detecting fine cracks. Zyglo was effective in detecting surface imperfections. Supersonic testing was effective in detecting cracks or voids too small to be found by X-Ray. An effort was made to correlate failures with the stress-rupture properties, damping capacity, and fatigue resistance. It seemed that the design and operating conditions influenced the nature of the failures. In the turbosuperchargers, for example, the buckets were more effectively damped than in the aircraft gas turbine buckets. Bucket bands in the turbosuperchargers touched and in some cases were upset by expansion. In the buckets without the damping effect, there is some correlation between failure in the service test and the high temperature fatigue strength.

Saldin and DeHuff (55) have said that the advantages of running bladed discs in an actual engine test are:

1. Identical conditions of normal engine operation may be duplicated.
2. No additional testing equipment is needed

There are howwver many disadvantages:

1. Optimum personnel safety is difficult and costly.
2. Costly engine damage occurs in case of premature failing of even a minor part
3. Rotational speed is limited by other engine parts
4. High operating costs
5. Temperature measurements are difficult
6. Design and configuration of the equipment is inflexible
7. There is little independent control of variables
8. There is difficulty in repeating test conditions and results
9. Measurement of creep is difficult without extensive engine teardown.

Stewart and Ellinghausen (13) have described the comparison of high temperature blade materials in a type B turbosupercharger. Twelve different alloys were tested as blade material. 142 blades of both wrought and precision cast alloy blades were mounted in the rotor disk of a type B turbosupercharger. Diesel oil was the fuel used. Test runs were made at 1200-1500 F for 50 to 150 hours. After each test run, the supercharger was removed from the test stand and taken into a constant temperature room where measurements were made to determine the permanent extension in the blades and rotor disk. Stewart and Ellinghausen (13) report what can happen in a service test as follows: " No new blades of high-strength alloys were available for replacement, and a set of vitallium blades was taken from a standard rotor for this purpose. This proved unfortunate as one of the replacement blades failed during the second 50 hour test run at 1500 F. — The failed blade segment damaged a considerable number of blades in the rotor. Forty-five of the damaged blades were removed and replaced with new vitallium blades.

Renewal of forty-five blades and damage to the reference circles made it necessary to reestablish the reference marks, renew the bearings, and balance the rotor."

An interesting point about the results of these tests was that two of the alloys which showed the maximum resistance to deformation at 1500 F in the laboratory creep tests stood approximately in the middle of the group tested in the supercharger. Thus it seems

that one of the major objectives in high temperature metallurgy is the development of a laboratory test for evaluating materials that will correlate with service experience.

Recently Hoffman and Ault (105) made a study of the application of statistical methods to the evaluation of gas-turbine blade failures. Routine laboratory tests are usually applied to newly available materials to rate them upon the basis of stress - rupture strength, creep rate, fatigue life and other properties. These data are valuable in the selection and development of materials in that they furnish a relatively fundamental basis for comparison and study. But, as has already been pointed out, the data are difficult to apply directly for service use because they fail to indicate the effects of complex conditions of stress, temperature, and atmosphere that exist in service. In order to approach service conditions, simulated service tests, simplified to reduce the number of variables and to improve control of variables are employed for evaluation of materials before use in service applications. If it were possible to correlate failures in the service test with failures in the laboratory test, it would be possible to eliminate much of the expensive service testing. For example, if a service failure can be attributed to fatigue or stress rupture, then the fatigue properties or stress rupture properties would be the important

criterion in the laboratory test.

Hoffman and Ault (105) determined the frequency distribution of time until failure of 142 cast cobalt base turbine blades mounted in each of two discs. In one wheel, the entire 142 blades failed. In the other wheel, 64 of the blades failed. The blades in the first wheel had an average life of 25.5 hours with a standard deviation of 11.5 hours. With this large standard deviation may be seen the fallacy of making very few tests to obtain data. The results of very few tests might place at one extreme or the other.

Reynolds, Freeman, and White (62) have said that erratic results in rupture tests at 1200 F on 19-9 DL were probably due to variation in sample location and partially to an inherent erratic characteristic associated with low deformations in the materials. They recommended sufficient tests to obtain a good average curve of stress against rupture time.

Mills (107) in discussing the design of a test coupon found by statistical analysis that when good steelmaking, forging, and heat treating practice had been established, there was as much variation in physical properties within any single forging as within an entire heat treatment batch. Tests from the end of any given forging then were no better criterion of the material in the center of that forging than were the tests from another forging of that heat treatment batch. Statistical methods for studying the properties of wrought steel products have been given in a series of papers from the Carnegie Institute of Technology.

Special equipment and techniques necessary in high temperature testing have been described previously. Further information may be found in references (13), (61), (71), and (106).

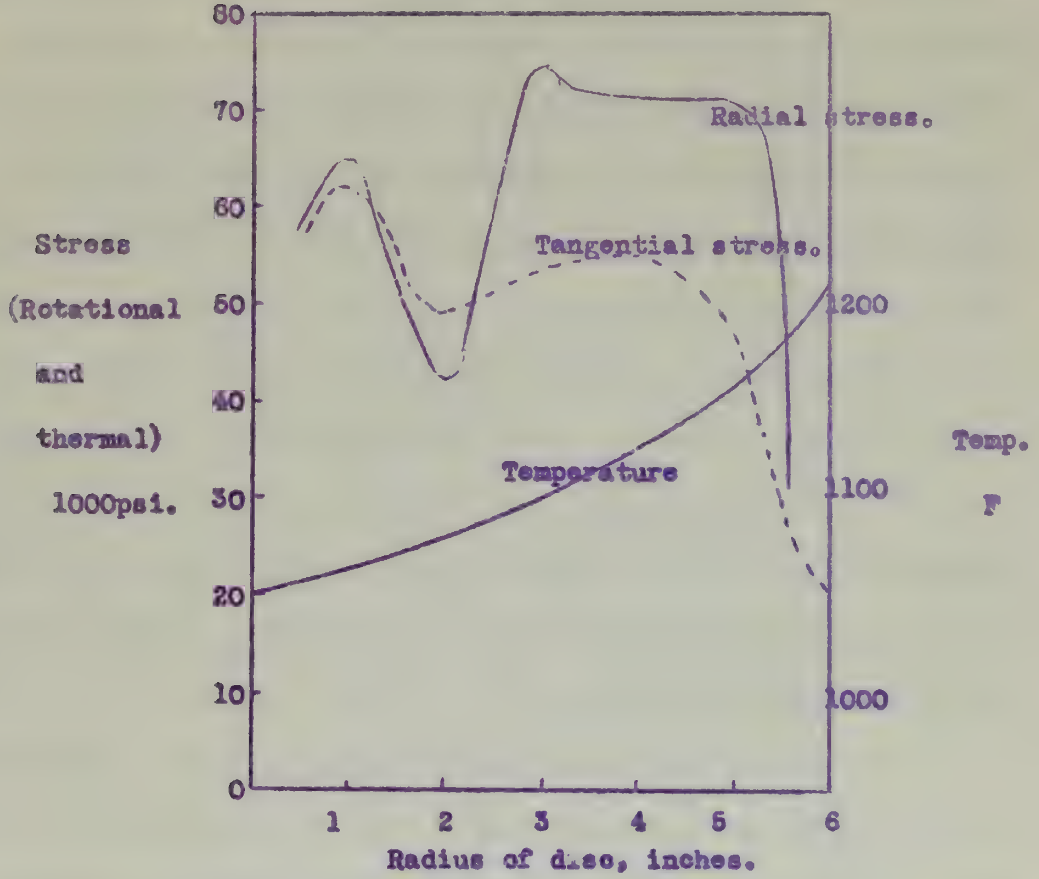

Figure 24. Stress and temperature distribution in a disc of 19-9DL rotating at 16,500 rpm. (55).

slip rings where strain gage signals were transmitted to a cathode ray oscilloscope.

The present designs of the Elliott gas turbine are based on the maximum stress theory of failure. The evaluation of the working stress is made on the basis of the stress - to rupture test in simple tension at elevated temperatures. On the basis of what has been said, it would appear that the design of the gas turbine is an undertaking involving many approximations.

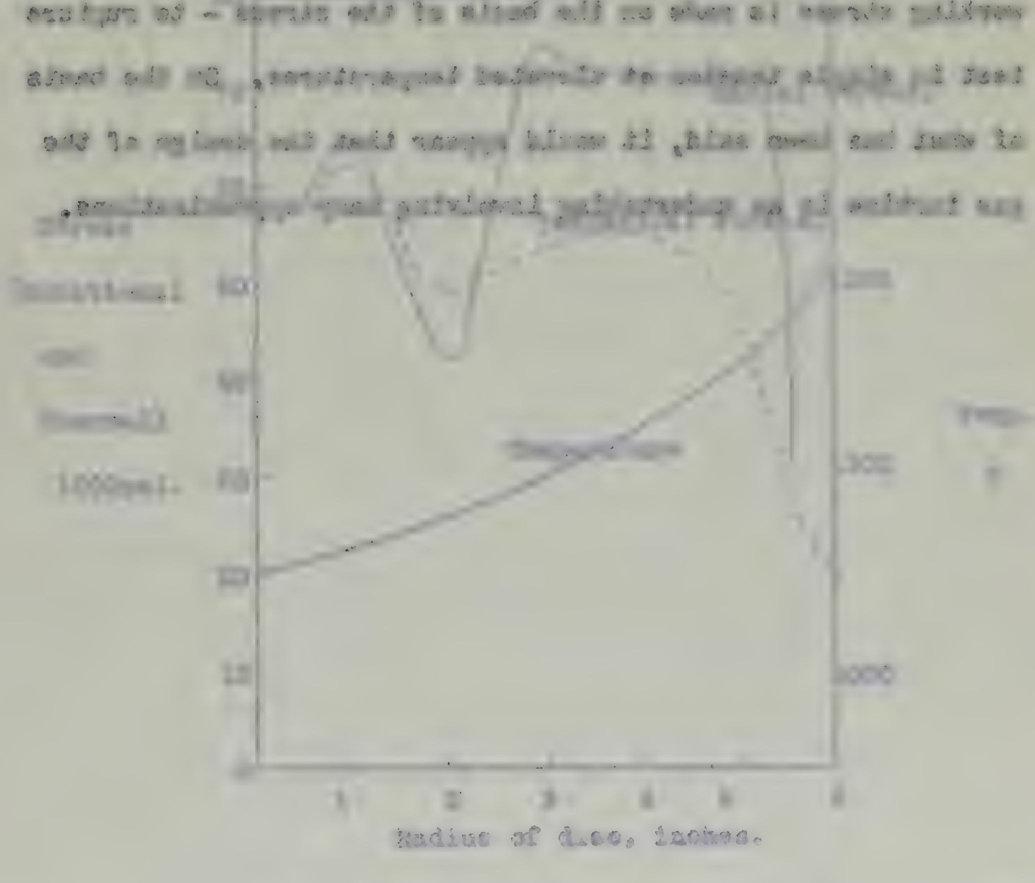

Figure 24, Stress and temperature distribution in a disc of 19-9DL rotating at 16,500 rpm. (55).

CHAPTER IX

STRESS AND DESIGN

In Figure 22 was shown the stress distribution in a blade. Figure 24 depicts the stress distribution in a disc with 150 F temperature differential from center to rim rotating at 16,500 rpm. As with the description of the stress distribution in the turbine blades, the stress problem in the disc is very complex. The disc is subjected to heat by conduction from the blades and may in some cases be directly heated by the combustion gases. The temperature differential existing between the center of the disc and the rim causes a differential expansion which introduces a system of thermal stresses that combine with the centrifugal stresses due to rotation (118) (122).

The elastic theory for disks operating at high temperatures for extended periods of time is completely inadequate. The accurate solution of the growth of turbine disks is an important and difficult problem. While the final stresses depend only on the state of creep ultimately reached, the growth depends on the entire cycle of life (123).

Difficulties connected with experimental stress analysis of discs and blades are great. (118) has described how high temperature wire-resistance strain gages were mounted at the bases of several blades and the vibratory stress of the blade was recorded during operation. Several components of the turbine were modified in order to provide passage of the wires from the strain gages to the

SUMMARY

A number of factors which must be considered in evaluating high temperature materials have been discussed. The little that is known and the great amount of unknown information concerned with these tests and properties has been pointed out. Warning was given at the outset that none of these factors can be considered individually since interaction and interdependence are important in the increase or decrease in the units of measurement expressing the quantitative measurement of the property. Stress, strain, time, and temperature must be considered simultaneously when evaluating materials at high temperature. This factor is the explanation why the obtaining of data is difficult and expensive.

Some rather obvious future requirements are:

1. Coordinated programs of research
2. Fundamental research on mechanism of creep
3. Studies of complex stress systems at high temperatures
4. Study of the effects of long periods of stressing at elevated temperatures on the stability of materials
5. Statistical methods of experimental analysis

In the interim, the designer of gas turbines must exercise careful judgment based on empirical relationships, adequate margins of safety, and on his ability to discern if a particular test or test method is applicable, reliable, and trustworthy.

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Alloy	C	Mn	Si	Cr	Ni	Co	Mo	W	Cb	Other
501	.11	[illegible]	.30	5			.5?			
S816	.31	.54	.26	19.1	19.7	44.4	2.??	4.0	3.0	3.5 Fe
4330	.31	.55	.11	.??	?.1					Bal Fe
S590	.47	1.33	.82	19.4	19.1	19.86	4.0	4.0	3.87	Bal Fe.
N155	.15	1.7	.51	21.3	19.0	19.2	2.9	2.1	1.1	
Vitallium	.20			27.5		Bal.		5.6		
61	.43	.??	.57	2?.2		Bal.				
[illegible]	.25	1.0		25		68	6.			
[illegible]	.35	1.0		26	52	55	6			
[illegible]	1.?	1.5	.5	20	30	20	3	2.2		2 Ta, Bal Fe
Hastel- loy B	.03	.90	.19		63.1		28.1			4.7 Fe
K42 B	.05	.70	.34	18	42	22				.51 Al, 2.6 Ti, 13 Fe
Inconel X	.09			15	73					.60 Al, 2.5 Ti.
Nimo- nic80	.04	.36	.47	21.2	74.2					
19-9DL	.3	1.0	.6	19	9		1.3	2	.5	.4 Ti,
Refract- aloy	.15	1.21	.29	20.6	20.7	31.4	8.4	4.5		13 Fe

Appendix I

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